1960

Investigation of heat-affected zone cracking in welded joints of modified HY-80 steel.

Henry, Eugene M.

Monterey, California: U.S. Naval Postgraduate School

http://hdl.handle.net/10945/12433

INVESTIGATION OF HEAT-AFFECTED ZONE CRACKING IN WELDED JOINTS OF MODIFIED HY-80 STEEL

EUGENE M. HENRY

UNITED STATES
NAVAL POSTGRADUATE SCHOOL

THESIS

INVESTIGATION OF

HEAT-AFFECTED ZONE CRACKING IN WELDED JOINTS

OF MODIFIED HY-80 STEEL

by

Eugene M. Henry

Lieutenant Commander, United States Navy

1960

INVESTIGATION OF

HEAT-AFFECTED ZONE CRACKING IN WELDED JOINTS

OF MODIFIED HY-80 STEEL

* * * * *

Eugene M. Henry

INVESTIGATION OF HEAT-AFFECTED ZONE CRACKING IN WELDED JOINTS OF MODIFIED HY-80 STEEL

by

Eugene M. Henry

Lieutenant Commander, United States Navy

Submitted in partial fulfillment of
the requirements for the degree of

MASTER OF SCIENCE

United States Naval Postgraduate School
Monterey, California

1 9 6 0

INVESTIGATION OF

HEAT-AFFECTED ZONE CRACKING IN WELDED JOINTS

OF MODIFIED HY-80 STEEL

by

Eugene M. Henry

This work is accepted as fulfilling

the thesis requirements for the degree of

MASTER OF SCIENCE

from the

United States Naval Postgraduate School

ABSTRACT

Heat-affected zone cracking in the welding of modified HY-80 steel (thicknesses greater than $1\frac{1}{4}$") has been a serious and costly problem in the nuclear submarine construction program.

Welded joints fabricated with a normal heat input, some with a preheating temperature of 150 F, some with no preheating, were examined microscopically and hardness, impact strength, and tensile strength measured. In these investigations attention was focussed primarily on the heat-affected zone. X-ray diffractometer tests were made to determine the presence of retained austenite in both the heat-affected zone and in the weld deposit. The specimens were fabricated by Mare Island Naval Shipyard under good shop conditions.

In welds made without preheating a considerable portion of the heat-affected zone, particularly the portion adjacent to the outer weld passes, showed low notch toughness, high hardness, and in a tensile test did not deform to any appreciable extent without cracking. In fact, the mechanical properties in this area are not greatly different from those of fully hardened base plate, and microscopic examination of this area revealed a high percentage of untempered martinsite. Spontaneous "delayed" cracking was observed in all incomplete welds which were fabricated without preheating. This cracking occurred both in the heat-affected zone and in the weld deposit.

Joints welded with a preheating temperature of 150 F showed somewhat better mechanical properties and considerably less cracking in the heat-affected zone. However, tensile tests of these welds, with the heat-affected zone oriented longitudinally with the specimen, revealed that cracks can and do occur in this zone with negligible elongation.

No retained austenite was detected either in the heat-affected zone or in the weld deposit.

Assuming proper welding technique, low hydrogen, and a normal heat input, the primary factors of heat-affected zone cracking appear to be (a) the high hardenability of modified HY-80 steel, (b) the rapid cooling rate provided by the massive structure, both in non-preheated welds and those made with a low preheating temperature, and (c) high restraint stresses.

The cooling rate in the vicinity of 550-600 F (low temperature cooling rate) appears to be an important consideration in this problem. Moderate temperature preheating has a marked effect on the rate of cooling in this temperature range, not only controlling the microstructural constituents formed in the heat-affected zone, but is helpful in minimizing the residual stresses occurring therein.

The higher alloy content plate is seemingly unweldable without preheating to a moderate temperature, probably greater than 150 F, if heat-affected zone cracking is to be controlled. Since it is not always practicable to preheat to such a temperature, it may be necessary to re-evaluate the relative importance of the yield strength and the weldability of the base plate.

The writer wishes to express his appreciation for the assistance and encouragement given him by Professors J. R. Clark and F. L. Coonan and for the generous assistance given by Mr. R. F. Edwards and Mr. A. Rasmussen, all of the U. S. Naval Postgraduate School. I wish also to acknowledge helpful information received from Professor E. R. Parker, University of California; Captain B. H. Andrews, Mr. L. Robbins, and Mr. H. G. MacKerrow, Mare Island Naval Shipyard; Mr. T. J. Dawson,

Ingalls Shipbuilding Corp; Commander E. C. Vicars, Bureau of Ships; and Lieutenant Commander K. N. Sargent, Office of Industrial Manager, 9th Naval District.

TABLE OF CONTENTS

GLOSSARY OF TERMS

Term	Meaning
Bead weld	A type of weld made by one passage of electrode or rod
Ductility transition temperature	The temperature corresponding to the onset of truly brittle behavior
Heat-affected zone	The portion of the base plate whose microstructure has been altered by the heat of welding
Low temperature cooling rate	Used in this paper and by some investigators to designate the rate of cooling in the vicinity of 600 F
"M_s"	The temperature where transformation to martensite begins
"M_f"	The temperature where transformation to martensite is complete
Modified HY-80	The higher alloy, heavier plate, of the two types of HY-80, nickel-chromium-molybdenum steel
Multiple pass welding	The process of depositing several successive layers of weld in the building up of a weld joint
"NDT"	Nil ductility temperature, generally used to designate the ductility transition temperature
Regular HY-80	The lower alloy, lighter plate, of the two types of HY-80 steel
"SSBN"	Ballistic missile nuclear submarine
Tee joint	A welded joint at the junction of two parts located approximately at right angles to each other
Unaffected zone	The portion of the base plate outside the heat-affected zone

LIST OF ILLUSTRATIONS

LIST OF TABLES IN APPENDIX I

SECTION 1

INTRODUCTION

THE NEEDS OF THE NAVY FOR A STEEL OF HIGH YIELD STRENGTH, NOTCH TOUGHNESS, AND GOOD WELDABILITY

With the increase of operating depths of submarines and of the destructive power of underwater nuclear weapons, the navy has an imperative need for a steel with optimum strength, toughness, and fabricability. Pressure vessels of modern steam plants, employing temperatures and pressures considerably higher than used in World War 2, and, more recently, atomic-reactor applications, also call for a steel which will furnish these properties with no additional weight.

The Bureau of Ships has specified the use of "HY-80", a high yield strength, highly notch tough, low alloy steel, for use in submarine pressure hulls and for an increasing number of critical applications. One disappointing factor in the use of HY-80 has been the high incidence of cracking in the welded joints of heavier plate steel, designated "modified HY-80", or "higher chemistry".

Alloy steels furnishing the needed strength and toughness have been fairly abundant for some time. However, one great limiting factor which has plagued industry for some time has been their poor weldability. This limitation is particularly serious in instances where proper preheating and postheating of the weld are not easily employed. Submarine pressure hulls, for example, where plate thickness is approximately 2", and where hull diameters reach 30', are extremely critical to weld, being complicated by the massive structure and high restraint stresses which are set up by the multiple-pass welding operation. Proper preheating and postheating are difficult, time consuming, and expensive.

COMPOSITION AND PROPERTIES OF MODIFIED HY-80 STEEL

HY-80 steel was first produced in 1947, to answer the need for a high strength, notch tough steel, having good fabricability. For heavier plate thicknesses (greater than 1½"), hardenability was increased by raising the nickel, chromium, and molybdenum content to provide good hardenability in thicknesses up to 6".

The designation "HY-80" steel includes both the "regular HY-80" and the "modified HY-80", the latter being of higher alloy content and limited to plates over 1½" in thickness. It is emphasized here that the term "HY-80" should not be used loosely--especially when referring to weldability. The higher hardenability of the modified HY-80 makes it much more difficult to weld, more susceptible to cracking in the heat-affected zone. This investigation deals exclusively with the modified HY-80 steel.

Modified HY-80 steel is a nickel-chromium-molybdenum steel with carbon content less than .23%. It is a fully-killed (deoxidized) steel, molybdenum being added to alleviate temper brittleness. Phosphorous and sulfur, elements which generally impair toughness and weldability, are kept low. Composition specification by elements is as follows:

C	Mn	P	S	Si	Ni	Cr	Mo
.23	.10	.04	.045	.12	2.68	1.29	.37
Max.	.4	Max.	Max.	.38	3.32	1.91	.63

The U. S. Steel Corporation has selected the following mid-range heat as an average composition for modified HY-80;

C	Mn	P	S	Si	Ni	Cr	Mo	Al
.13	.16	.009	.013	.10	3.08	1.76	.49	.025

HY-80 steel is highly notch tough and completely resistant to brittle fracture at temperatures of -20 F and higher. Its yield strength is equal to or greater than 80,000 psi.

The Bureau of Ships has specified that the Charpy V Notch impact energy of modified HY-80 steel be 30 ft-lbs. at -120 F for plate

1 3/8" thickness or greater. The impact energy has a fairly linear rise to approximately 120 ft-lbs. at 72 F.

Typical mechanical properties of quenched and tempered modified HY-80 base plate of 3 3/8" thickness (Heat 68M549) are tabulated below:

YIELD STRENGTH AT .2% OFFSET, PSI	77,800
ULTIMATE TENSILE STRENGTH, PSI	105,800
ELONGATION IN 2 IN. GAGE LENGTH, %	25
REDUCTION IN AREA, %	73
HARDNESS, BHN	216

The weld metal specified by the Bureau of Ships for manual arc welding is type 11018 electrode, MIL-E-19322. Its composition is specified as:

C	Mn	Si	P	S	Cr	Ni	Mo	Va
.10 Max	1.3- 1.8	.6 Max	.03 Max	.03 Max	.4 Max	1.35 2.5	.3- .55	.05 Max

Type 11018 electrode has a low hydrogen, iron powder type coating.

PROBLEMS IN WELDING MODIFIED HY-80

Difficulty in welding modified HY-80 steel was encountered early in the nuclear submarine construction program. At this time many questions were unanswered as to the best or optimum

(1) amount of preheating and postheating required

(2) amount of heat input in welding

(3) type of electrode to use

(4) welding method and technique.

The serious nature of the problem is pointed up by representative statements of welding engineers and ship-builders who represented their respective activities at the Bureau of Ships Conference on "HY-80 Steel Fabrication in Submarine Construction," in July, 1958. Some typical comments are quoted as follows:

In fact we find it practically as critical to weld as the more familiar STS material. T. J. DAWSON, Ingalls Shipbuilding Corporation.

We realized that our difficulties were serious, since some portion of all "tee" welding to 1 3/8" thick HY-80 material showed toe cracking. Our experience indicates that toe cracks submerged in the hard zone may later propagate to the surface. R. H. CUNNINGHAM, Newport News Shipbuilding and Drydock Company.

We have made several transverse "tee" bend tests and find that failure universally occurs in the hard zone. R. H. CUNNINGHAM, Newport News Shipbuilding and Drydock Company.

It has been found that toe cracking in HY-80 propagates through the grain coarsened heat-affected zone. E. H. FRANKS, Philadelphia Naval Shipyard.

Probes revealed that the transverse cracks in weld metal (in welding internal frames) originated in the heat-affected zone. The number and locations of the cracks made it mandatory that we remove all the weld and reweld the entire joint (hull plate, 100 ft. circumferential butt). S. I. ROBERTS and C. E. COLE, Portsmouth Naval Shipyard.

After extensive consultation with welding engineers and materials test laboratory personnel from Mare Island Naval Shipyard, and after studying the contents of Reference (3), the following preliminary conclusions were drawn:

(1) The heat-affected zone appears to be the area of highest incidence of cracking.

(2) Cracking is more pronounced in modified HY-80, the higher alloy content steel.

(3) Cracks occur in "tee" joints to a greater extent than in butt joints.

SECTION 2

FACTORS AND CONDITIONS WHICH PROMOTE CRACKING IN WELDED JOINTS

GENERAL

In the great majority of cases, cracks in the welded joints of structural steels are due to stresses set up by three independent factors: mechanical constraint, volumetric phase change in the steel, (causing high localized stresses), and entrapped hydrogen [17]. These three factors are discussed in this section, plus related matters which, in some way, influence the susceptibility to cracking.

Experience with a wide range of alloy steels has shown that if the maximum heat-affected zone hardness is not greater than 350 Diamond Pyramid Hardness, the steel generally remains crack-free under most welding conditions [1, 17], i.e., if volumetric changes are small enough, the effects of the other factors contributing to cracking are minimized. This criterion is recommended only as a general guide. When maximum hardness exceeds 350 Diamond Pyramid Hardness, preheating is strongly recommended.

In this investigation, the maximum hardness readings in the heat-affected zone were: Knoop Hardness Number 508 for welds made without preheating and Knoop Hardness Number 452 for welds preheated to 150 F. These hardness readings are comparable to Diamond Pyramid Hardnesses of 485 and 435 respectively.

The maximum hardness rule of thumb indicates that trouble can be anticipated when welding modified HY-80 without preheating, and when preheating to only 150 F.

LOCALIZED STRESSES

The welded joint is always accompanied by residual stresses. Residual stresses are defined as those which are present in a body without any external forces acting.

Longitudinal tensile stresses occur due to contraction of the weld deposit upon cooling, reaching a magnitude of the yield strength of the base metal. The adjacent base plate attempts to restrain the contraction and, as a result, is subjected to this same stress of high magnitude.

Transverse stresses due to solidification of weld metal in the weld joint are also present, but are of much less magnitude than are the longitudinal stresses. Figure 9.4 of Reference 10 depicts these stresses.

Thickness direction stresses in welded joints have been detected and evaluated [10]. Certain welding procedures produced triaxial tension stresses at the center of the plate in butt-welded joints, both in 2" and 3/4" steel plate, particularly pronounced when using "hand, stringer bead" welding technique. Thickness direction stresses were tabulated as high as 49,000 psi in tension.

Residual stresses due to change in crystal structure of the steel upon heating or cooling are very significant, and, since a volumetric change occurs in the crystal, are often a principal cause of quench cracking. "Delayed cracking" is often attributed to this transformation.

Parker [10] gives instances of spontaneous cracking in welded joints before any load was applied, which he attributes to residual welding stresses. He indicates that a low temperature stress relief anneal after welding greatly alleviates such a tendency.

It is impossible to evaluate accurately the combined effect of these stress systems. In spite of numerous investigations, little is

known about the effect of residual stresses on failures. Low temperature brittle fractures in steels containing crack starters are certainly greatly affected by residual stress.

Temperature plays an important part in fracture under residual stress. If the temperature is above the ductility transition temperature[1] of the metal in question, fracture occurs at a very high applied stress, and residual stress seems to have little effect on the fracture stress. If the temperature is below the ductility transition temperature, highly stressed areas containing a notch can and do fracture at low applied stresses, or, in some cases, even with no applied stress (Figure 7 - 9). If the temperature is considerably below the transition temperature, the fracture will propagate to complete failure unless the crack reaches an area which is considerably warmer, or one in which the applied stress is very low, or where a compressive stress exists [7, 10].

It is believed that residual stresses have little or no effect when fracture occurs in a ductile manner. However, it should be borne in mind that a steel failing in a ductile manner can easily lose that ductility in the heat-affected zone of a weld. This is often the case with alloy steels.

Thermal stresses are a function of thermal expansion and thermal conductivity. High local stresses develop when a steel has a high coefficient of thermal expansion and low thermal conductivity. Two metals to be joined should have approximately the same thermal expansion coefficient. Thermal conductivity lessens with alloying--hence there is a greater susceptibility to thermal stresses, which have brought about failures.

---

[1]the temperature corresponding to the onset of truly brittle behavior, or where ductility is nil.

Two phenomena, "thermal shock," and "thermal fatigue," are deleterious to metals and should be avoided when possible. Thermal shock is the condition arising from a rapid application and removal of heat. Thermal fatigue is brought about by thermal cycling, or a large number of thermal shocks.

Austenitic steel has a greater linear thermal expansion coefficient than does ferritic. Therefore, a steel containing both austenite and ferrite would develop local thermal stresses when heated. Residual stresses in welded joints can be alleviated to a great extent by controlling the rate of cooling and by stress-relief annealing. Thermal stresses must be dealt with by the proper selection of materials, and avoiding, as much as possible, thermal shock and thermal fatigue.

Welding engineers and welders know the importance of good joint design in order to alleviate high local stresses which accompany sharp corners, small radii, improper fillets, etc. The "T" joint contains stresses in the fillet weld markedly higher than does the butt joint; therefore is more susceptible to cracking. Proper contouring of welds is of great importance. In fact, the increase of weld joint reliability by proper joint design completely overshadows the effect of a considerable change in composition of the metal.

RESTRAINT

Restraint of a welded joint is a major factor in weld cracking. Highly restrained plates are generally acknowledged to be more difficult to weld successfully. Massive welds usually contain very high restraint stresses. These stresses interact with the ever-present localized stresses of welding to increase the tendency of the joint to fail.

Investigators are in general agreement that at high ambient temperatures the stresses causing fracture are due primarily to restraint; at temperatures below the ductility transformation temperature the residual stresses play a dominant role, and fracture can occur with very low restraint stresses.

If, upon cooling, there is retained austenite in the welded joint, restraint stress will assist the initiation of cracking by inducing transformation of the austenite to martensite.

It is important to keep in mind that when restraint stresses are high, as in pressure hull welds, an inclusion, notch, small crack, etc. would effectively concentrate the stress in that area, the local stress in some cases equals the yield point. For example, at Mare Island Naval Shipyard, strain-gauge measurements around a circumferential butt weld revealed stresses as high as 60,000 psi. There is little doubt in this case that in some areas such as crack extremities, inclusions, porosity, etc., the stress level approaches or equals the yield point of the metal.

HYDROGEN EMBRITTLEMENT AND EFFECTS OF OTHER GASES

Embrittlement of welds may ensue locally in low and medium strength steels by molecular hydrogen entrapped in solidifying metal. This embrittlement is more general and more severe in high strength welds [1].

The solubility of hydrogen in molten steel is approximately .005 percent by weight of gas. Its source may be the atmosphere, the electrode coating, the metal core of the electrode, or the parent metal itself.

The phenomenon is explained as follows: During welding the hydrogen dissolves in the weld pool, and penetrates the heat-affected zone of the base metal. On solidification, most of the hydrogen is ejected, but some becomes entrapped in the rapidly solidifying metal in the form of

blow-holes. Some authorities say that the effects of the blow-holes are further harmful in that the gas is retained under high pressure and creates high residual stresses.

Hydrogen embrittlement is minimized by use of low hydrogen type electrodes, welding in an inert atmosphere, keeping moisture to a minimum by baking the electrodes and carefully cleaning the joint to be welded, and protection of welded area from the weather.

Nitrogen tends to embrittle and to decrease ductility. It may be in the form of iron nitride needles, in solution in steel, or gas entrapped in cavities. Its source is normally either from the steel itself or from the atmosphere. Rapid cooling may retain nitrogen in solid solution. Upon aging or subsequent heating, nitrogen is found to precipitate from solution in the form of iron nitride.

Oxygen embrittles mild steel and may cause grain boundary cracking. The effect of oxygen on commercial steels is not well known or understood. Oxides do lower the notch ductility even when the fracture is 100% shear.

Exclusion of air from the molten metal and cleanliness and dryness of the joint will greatly decrease the susceptibility to brittleness from these latter two gases.

Nitrogen and oxygen may contribute to porosity, although the majority of porosity is attributed to hydrogen and other gases.

Carbon monoxide is one of the most common sources of blow-holes in steel welds. Excessive sulfur is a source of porosity, due probably to hydrogen sulfide gas.

Porosity, besides decreasing the effective cross-section of the metal, has the more dangerous "notch effect" in concentration of stresses, frequently acting as the source of cracks in the heat-affected zone or in the weld deposit.

THE COOLING RATE AND HEAT-AFFECTED ZONE HARDNESS

The narrow band of base metal adjacent to the weld deposit so affected as to show microstructural change is referred to as the heat-affected zone. Metal immediately adjacent to the weld is heated to the melting point in a few seconds. The unfused metal near the weld is subjected to a variety of temperatures and cooling rates.

Since structure largely determines mechanical properties, the heat-affected zone should have a microstructure which provides the best attainable toughness with required strength. If the heat-affected zone is martensitic, it will crack at a very low strain and this crack may propagate by cleavage, causing the weld to fail in a brittle manner [1, 10, 17]. The extent of formation of martensite is directly controlled by the cooling rate.

In the steels of high hardenability, such as modified HY-80, the cooling rate at relatively low temperatures appears to be of prime importance in that it affects the toughness of the welded joint. The toughness is severely impaired when this rate is too high [6]. The cooling rate of a welded joint depends on the plate temperature, the thickness of the plate, heat input, and joint geometry. The effect and importance of the low temperature cooling rate is discussed in Section 4.

Preheating the parent plate before welding is perhaps the simplest and most effective way in most steels to minimize weld cracking. A relatively low temperature, 200-400 F is surprisingly effective, with higher temperatures required as carbon and alloy content increase [5, 10].

The cooling rate may also be controlled by immediate postheating the weld.

The size of the section (or thickness of the base plate) is a considerable factor affecting the cooling rate. Thick plate structures

furnish a drastic quench of the weld deposit and heat-affected zone, and are usually accompanied by high restraint stresses. "T" joints have a stronger quenching action than do butt joints simply because of the greater mass.

If alloying elements are present in the metal being welded, the cooling rate necessary to produce high hardness in the heat-affected zone is normally less than in plain carbon steels. Hence, hardness and consequent brittleness is more difficult to avoid during the welding cycle.

HARDENABILITY

Hardenability is the measure of the degree to which a steel can depth-harden under certain conditions of heating and cooling. Another way of describing hardenability is the ease with which a steel can harden fully at a relatively low cooling rate.

Carbon content is essentially the sole factor in determining the hardness attainable. However, the depth to which this hardness is attained for a specified cooling rate is determined not only by carbon, but by certain alloying elements as well. Therefore, additional alloying elements, as well as carbon, will increase the width of the hardened area in the heat-affected zone.

Manganese, chromium, nickel, molybdenum, copper, and tungsten, either singly or in combination, tend to increase the hardenability, thus the difficulty in welding a steel.

Essentially, the high hardenability of alloy steels is the consequence of the fact that the austenitic solid-solution formed at high temperatures remains intact during the early stages of cooling, and at lower temperatures transforms to a hard martensitic structure since structure formed during the transformation depends upon transformation temperatures.

Two principal factors can be employed in accounting for the stability of the austenitic solid-solution. First, the alloying elements lower the temperature at which martensite begins to form. Secondly, the ease of formation of a nucleus, ferrite or carbide, necessary for non-martensitic transformation, decreases as the rate of diffusion of the carbon atoms decreases. The diffusion rate, being a function of temperature and composition, has been lowered.

In general, any local hardness of the steel within the heat-affected zone is accompanied by a corresponding decrease in ductility. The occurrence of this hardening is often associated with cracking. If the hardness is too great, the affected zone is unable to deform plastically to accommodate the very high contractural stresses set up in welding--principally the tensile stress which is greatest in the direction longitudinal with the weld, and cracking must occur. This danger is increased as the hardened band is broadened and made more continuous by increasing alloy content.

NON-METALLIC INCLUSIONS

Non-metallic inclusions play an important part as crack starters in welded joints. They not only decrease the tensile and yield strength of the metal, but they act as "notches" in their capacity as crack starters in hardened zones. These inclusions are found both in the base metal and in the weld deposit.

The most common types of non-metallic inclusions are: slag, sulphides, dirt, dross from ingot mold, and oxidized splashings (found in base metal, developing while pouring the ingot). All of these are objectionable, since they are both weak and brittle and create objectionable discontinuities in the metal.

The effect of holes drilled in steel plate is well known in that they create high local stresses and decrease the cross sectional area of the plate. An analogy may be drawn between these holes and non-metallic inclusions. The holes must be thought of as very close together, and if the holes are jagged in shape the analogy is even better.

The objectionable nature of inclusions is multiplied if they occur in a continuous pattern or network.

Large carbide particles widely separated in a ferrite matrix permit a dangerous pile-up of dislocations in that the inter-particle path is relatively long as compared to that in an area of many small, well-dispersed particles. In the former case, a crack may easily develop at the interface of the large carbide particle.

When the carbides are uniformly distributed through the structure, the effect mentioned above does not occur. The tiny particles of carbide break up the ferrite more favorably. Upon stressing, the metal undergoes a uniform distortion, with no local stress concentrations or "pile-ups" of dislocations [10]. In a steel with uniformly distributed carbides, little flow takes place in any one grain before it is checked by the resistance which is offered by a globule of carbide. Flow then starts at a point in some other area and goes on until it is checked in its turn. Consequently, the flow is uniformly distributed in the different grains, greatly decreasing the tendency of the steel to fracture.

Higher carbon steels, commonly used when strength is needed, are susceptible to brittleness and cracking on rapid quenching. This is largely due to the carbon being entrapped in a strained lattice, not having had time to diffuse out on cooling.

Strong carbide formers such as titanium, columbium, and tantalum, put carbon into an "inactive" form and make the steel behave on quenching

as if it had lower carbon, but the undesolved carbides give a high "hardness" as measured in terms of wear resistance.

RETAINED AUSTENITE

As steel is heated from room temperature to above the upper critical temperature, it transforms entirely to austenite, a phase not found at room temperature under equilibrium conditions.

In the relatively deep-hardening steels some austenite is generally retained in the heat-affected zone after the weld has cooled to room temperature. The amount of austenite varies with carbon and alloy content. It is accepted that austenite is not normally detrimental in a steel at room temperature, but, beyond a small percentage, it is detrimental if in time it transforms to stable phases, undergoing a volumetric change, greatly increasing local stresses and causing cracks to form.

Measurable quantities of austenite were found in a series of plain carbon steels containing from 1.07% down to .20% carbon. However, in steels with less than .5% carbon, relatively little austenite was subsequently transformed by refrigeration. In this particular work, retained austenite was visible under the microscope (at 1800X) only when the austenite content exceeded about 10%, which occurred in the higher carbon steels. The austenite decreased rapidly with decreasing carbon content, with .4% austenite measured in .2% carbon steels [18].

GRAIN SIZE

Large, coarse, fairly geometric crystals, brought about by heating to a very high temperature, are objectionable in any metal that is to be used for strong structural components. It is generally acknowledged as being weaker and less notch-tough. Dislocations can easily move through such a structure.

Coarse austenite grains affect many other properties, promoting greater depth of hardening, more retained austenite, higher internal stresses on quenching, and more embrittlement by cold working. Other effects, not directly connected with welding, are not mentioned here.

Grain coarsening is a function of time at high temperature, and to a certain extent, is controllable in a welded joint. Very high energy inputs will produce extensive grain coarsening in the heat-affected zone. Heat input may be varied by technique and method of welding. Certain grain growth inhibitors such as titanium, vanadium and zirconium promote fine-grained characteristics in steel.

Cracking in martensite, which forms extensively in the heat-affected zone is shown to increase with grain size [3, 4]. "Toe cracking," forming in the heat-affected zone at the edge of welds, is reported extensively in the use of modified HY-80 steel in new submarine construction. These toe cracks propagate through coarse grain areas, stopping in the fine grain areas [3].

THE WELDING ELECTRODE

Normal analysis of chemical composition does not assure welding properties. However, the composition of the weld metal should approach that of the base metal in order to avoid discontinuities in physical and chemical characteristics of the weld. The only real test of a satisfactory weld rod is the result of the actual welding operation and the mechanical properties of the weld deposit individually and as a part of the weld.

Electrodes with high or moderately high hardenability are often used for structural steels when some means is available for postheating. If the weld metal possesses sufficient hardenability, and if restraint

stresses are high, cracking will initiate within the weld deposit, being triggered by the ever-present inclusions, porosity, cratering, etc. Too, should cracking originate in the heat-affected zone, it propagates easily through such a weld deposit.

OTHER CONSIDERATIONS

Homogeneity is, of course, desirable. Inhomogeneity is generally found in the weld deposit with its dendritic or columnar structure, depending on the cooling rate. Inhomogeneity will manifest itself in areas alternately high and low in carbides, in some cases creating highly decarburized, softer, and much weaker areas. Weld deposit cracks have been observed to follow the boundary, separating areas of alternately high and low carbon content. This is observed in type 11018 electrode deposit used with HY-80 steel.

Sulfur when not combined in MnS may combine with iron to embrittle the steel. Ample manganese normally insures against this embrittlement. MnS has recently been discovered in the grain boundary of an embrittled low alloy steel [19]. Although the MnS was identified, insufficient evidence is available to attribute the embrittlement to it. However, this work suggests that MnS may be more important in welding failures than has been considered in the past, and further work would be necessary to determine its importance.

SECTION 3

BRITTLE FRACTURE

BRITTLENESS DEFINED

A brittle area literally means one in which failure occurs with very little or no plastic deformation. Brittle fractures have ambiguously been called cleavage, crystalline, or granular type fractures. The words "crystalline" and "granular" should be reserved, however, for describing fracture appearance. "Cleavage" describes a mode of fracture (cleavage plane), and does not always indicate a brittle failure.

Parker [10] states that the path of fracture at ordinary temperatures is normally across the grain--not intergranular. In brittle fracture, failure occurs across cleavage planes (but not all cleavage plane fractures are necessarily due to brittleness). Ductile fractures, however, take place along the slip plane. In steels intercrystalline, or grain boundary, failure is less common than transgranular failure, occurring in such instances as (1) prolonged loading at elevated temperatures, (2) "caustic embrittlement," or (3) in some temper-embrittled low-alloy steels.

Relating the nature of the stress to the planes in the crystal, failure by cleavage is caused by normal (tensile) stresses, whereas shear failure is promoted by shear stresses. In a polycrystalline material cleavage fracture appears bright and granular, whereas shear failure appears silky gray and fibrous. Cleavage fracture is commonly called "granular." Shear is referred to as "fibrous."

Brittle fracture is characterized by negligible lateral contraction at fracture (of the order of 1%), the absence of "shear lip," and a bright and "granular" fracture surface.

BRITTLE CRACK INITIATION

Moderate success has been achieved in explaining brittle fracture. Extensive tests in recent years have shown that brittle failures originate at some discontinuity or defect due to design, fabrication, or repair. These include corners, notches, cracks, arc strikes, slag inclusions, undercutting, weld craters, etc.

The conditions sufficient for the initiation of brittle cracks in steel under static loading [7, 10, 16] are enumerated as follows:

a. There must be a stress raiser such as is mentioned above.

b. The temperature must be below the ductility transition temperature of the steel.

c. The stress raiser must be in an area of high tensile stress (combining residual and applied stress).

It is emphasized that the ductility transition temperature of the unaffected base plate may have very little significance in relation to the brittle fracture occurring within the heat-affected zone, where a completely different microstructure exists. In fact, the heat-affected zone may be susceptible to brittle fracture at room temperature while the unaffected base plate has an impressive toughness at very low temperatures.

Several factors may contribute to initiation of brittle fracture. When the total effects of contractional stresses due to welding, high restraint stresses, prestrain due to forming, and high hardenability of the steel are combined, brittle fracture may occur, or a potentially dangerous area may exist ready to fracture when the temperature has dropped sufficiently.

Experiments by KIHARA and MASUBUCHI [7] with a low notch-tough low carbon steel indicate that if there exists a sharp notch, such as

a weld crack, inclusion, lack of fusion, etc. in the highly stressed zone, the high residual stress in a welded joint may act as a trigger to initiate a brittle fracture on the application of even a low applied stress. Work by the British Welding Research Association points even more strongly to the residual stress as the origin of low stress brittle fracture.

Parker [10] has shown that triaxial tension stresses exist at the center of the plate in butt-welded joints. The thickness direction stress is substantial. This can constitute a further contribution to the initiation of cracking at low applied stress levels.

A general conclusion from crack re-initiation tests is that re-initiation of a brittle crack which has once terminated does not occur until the average applied stress reaches a value as high as the yield stress [7, 10, 15]. This would indicate that local residual stresses have been greatly alleviated by the cracking, and the residual stress has no effect on the fracture stress at the time of re-initiation of the crack.

BRITTLE CRACK PROPAGATION

Considerable work has been done recently in the field of crack propagation. Many of the conclusions are tentative, but some few points seem to be well established from reliable data. Once a brittle crack is started, it is generally concluded that for the crack to propagate spontaneously, these conditions must exist concurrently [7, 10, 15].

a. The temperature must be below the ductility transition temperature of the steel.

b. The average applied stress need only exceed a nominal value of about 10,000 psi.

Frequently, a brittle crack will start then suddenly terminate, having alleviated localized stresses, when either (1) the average applied stress is less than the nominal 10,000 psi or (2) the temperature is above the ductility transition temperature.

STRAIN-AGING

Several high strength low alloy steels have been shown to be susceptible to embrittlement by strain-aging [8]. This susceptibility to embrittlement from cold working and subsequent heating can be reduced, or even eliminated, either by the presence of suitable components added to the steel during manufacture, or by refining the grain of the metal [1].

Work by Rubin, Gross, and Stout [8] on the effects of heat treatment on heavy-section pressure vessel steels demonstrated that normalized and drawn HY-80 type steel is subject to embrittlement by strain-aging. Even though this steel suffered no harmful loss in elongation and reduction in area on strain-aging, it showed definite brittleness, the 15 ft/lb Charpy V-Notch impact energy transition temperature rising to 24 F after specimens had been pre-strained then aged at 700 F. All the pressure vessel steels tested in this work were susceptible to strain-aging. HY-80 showed the greatest loss in notch toughness and the greatest increase in tensile strength (both at 700 F) of the nine steels tested. Standard stress-relief temperatures for these steels obliterated practically all of the embrittlement induced by pre-straining.

BRITTLE FAILURES CORRELATED TO NOTCH TOUGHNESS

Brittle fracture service failures are correlated with laboratory notch-toughness tests in work done by Puzak, Babecki, and Pellini [14]. The ductility transition temperatures, determined by drop-weight tests, were demonstrated to correlate with the service failures of all materials

which were investigated. The failures occurred at temperatures below the ductility transformation temperature in each case.

Many investigators are now in agreement that the best criterion for evaluating the probability of a steel to fail in a brittle manner is the ductility transformation temperature as found by impact testing [10, 14].

SECTION 4

EFFECTS OF PREHEATING, POSTHEATING, AND OTHER THERMAL EFFECTS

PREHEATING

The effects of preheating in welding can be summarized as follows:

a. It reduces the effects of thermal gradients, hence thermal contraction cracking of structures under constraint.

b. It decreases cooling rate, especially at lower temperatures, decreasing the amount of martinsite formed.

c. It lessens residual stresses due to phase change since the residual stress, unable to exceed the yield stress, is less at a higher transformation temperature.

d. The outward diffusion of hydrogen on cooling is assisted, thus reducing risk of high concentrations and cracking.

Experience has shown the desirability of preheating in welding any steel of relatively high hardenability. Increasing alloy content normally requires a higher preheating temperature if toughness is important in the welded joint.

Considerable attention has been given recently to the low temperature cooling rate of a welded joint. A relatively low preheating temperature has a great effect on this cooling rate. For example: Franks [3], experimenting with heat-affected zone cooling rates versus preheating temperatures, demonstrated that the heat-affected zone of a nickel-chromium steel, 1" plate, preheated to 70 F cooled approximately eight times as fast in the range 550-600 F as it did when preheated to 300 F. This great difference is very significant if one recalls the work done by Berry and Allan [6] relating low temperature cooling rate to heat-affected zone cracking in low-alloy steels. They found that for several

low alloy steels the number of heat-affected zone cracks decreased as the low temperature cooling rate decreased.

In examining the effect of preheating on the cooling rate of the heat-affected zone, little effect is noted at higher temperatures. However, once the temperature has dropped to say 600 F, a marked decrease of the cooling rate occurs when preheating to 300 F as compared to the cooling rate when preheating to 70 F. In a steel of high hardenability and with a relatively high Ms temperature, such as modified HY-80, the pronounced decrease in cooling rate in the vicinity of the Ms would slow the transformation and deter the formation of some martensite in favor of a more ductile low temperature bainite. Residual stresses would be greatly decreased, thereby lessening the susceptibility to cracking in the heat-affected zone and in the weld deposit.

Figure 4-1, reproduced from Reference 3, illustrates the marked effect of a moderate preheating temperature on low temperature cooling. Figure 4-2 superimposes hypothetical cooling rate curves on isothermal transformation temperature curves for modified HY-80 steel, indicating the probable effects of preheating in alleviating high residual stresses and possibly decreasing the extent of martensite formation.

POSTHEATING

Tempering is the reheating of hardened steels to some temperature below the critical range for the purpose of toughening the material. A considerable increase in toughness can usually be attained with only a slight sacrifice of hardness and strength.

Low temperature stress-relief annealing is principally for the relief of gross internal stresses in the welded joint, and is normally effective in increasing the toughness of the material without necessarily

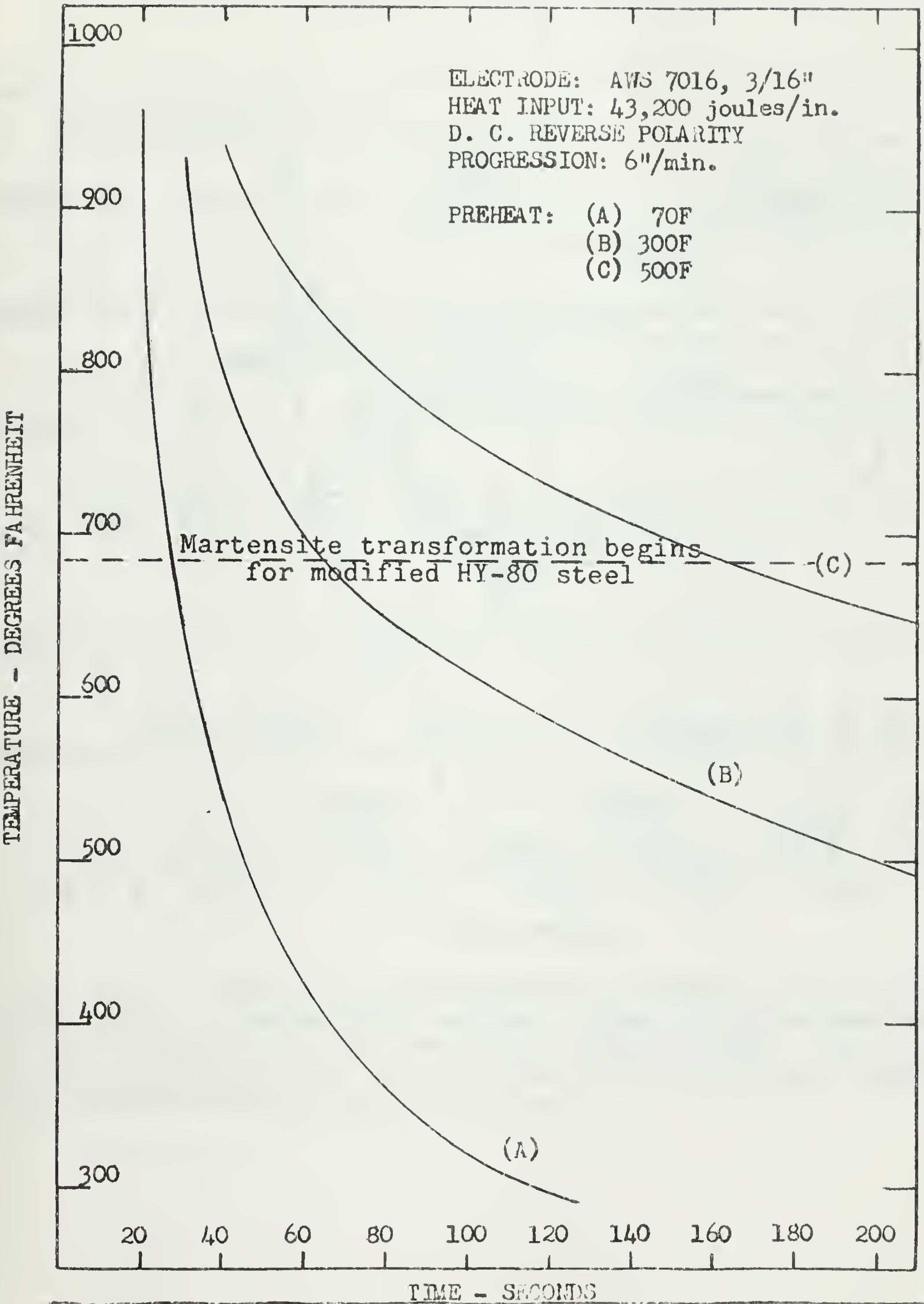


Figure 4-1. Low temperature cooling rate curves for the heat-affected zone of a nickel-chromium alloy steel. Note the marked difference in cooling rates in the vicinity of 600 F between (A) and (B).

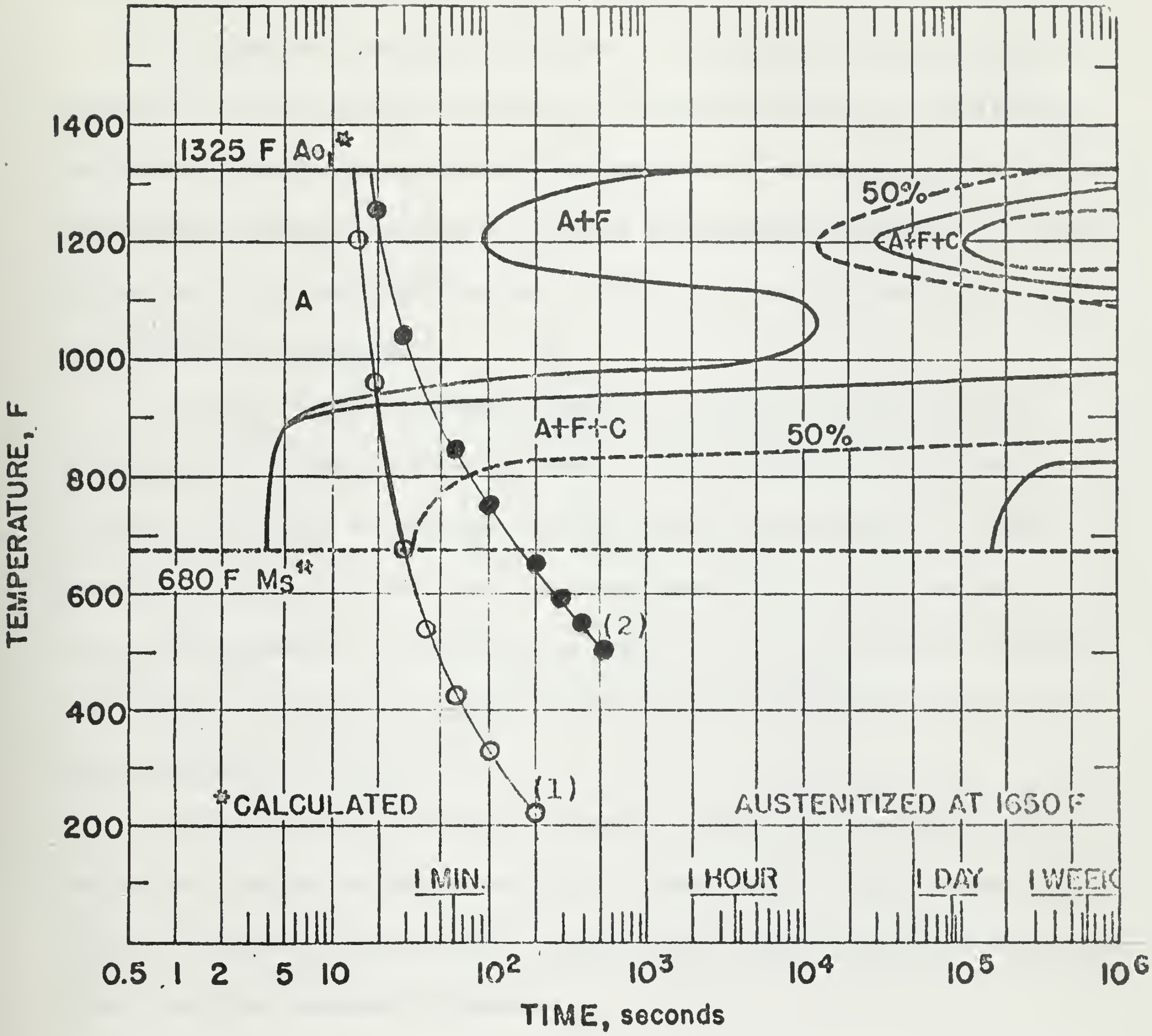


Figure 4-2. Isothermal-transformation diagram for modified HY-80 steel. Hypothetical cooling curves are superimposed for low (1) and moderate (2) preheating temperatures. Note that for the slower cooling rate less martensite is formed and its early transformation is more gradual.

affecting the microstructure to a great extent, since the temperature and duration are low and relatively short.

In tempering, the steel changes to a controllable extent from an unstable, or metastable, martensitic condition toward an equilibrium condition, a mixture of ferrite and cementite. In prolonged tempering, martensite gradually decomposes by the rejection of carbides from solid solution, the structure changing to ferrite in which fine particles of cementite are dispersed.

In a steel which has been tempered to contain many fine, well-dispersed particles of iron carbide, a crack starting at one side of a crystal is offered no straight and easy path along which to travel in order to reach the other side. As mentioned before, the carbide particles, strong and hard, block its movement. Too, the number of dislocations which can pile up in one localized area is limited by the small inter-particle spacing [10].

Steels differ in the heat treatment required to restore desirable mechanical properties after welding. Steels with a high hardenability normally require stress-relief annealing to insure a tough welded joint, free from high residual stresses.

Stress relief annealing is not always practical, especially for massive welds such as in submarine hulls. In this case, fabricators often must rely on preheating alone for controlling residual stresses and microstructure.

MULTI-PASS WELDING AND ITS TEMPERING EFFECT

In ship structural work, most welds made in heavy plate are multi-pass welds. They are built up by depositing several layers of weld-metal, one upon the other, in separate, successive operations. The

upper weld passes are generally beneficial, effecting some grain refinement and, to some extent, tempering both the underlying weld deposits and lower portions of the heat-affected zone. Often this effect is utilized in welds which are not to be heat-treated after welding, but it is not always effective in providing an adequate temper for steels with high hardenability.

HEAT INPUT

Heat input in welding is an important consideration, and is a major factor in the success or failure of a welded joint. Too great a heat input promotes undesirable grain growth in the heat-affected zone. Too little heat may cause poor penetration, inadequate diffusion at the weld junction, and in some cases, permit an excessive rate of cooling if there is danger of over-hardening the heat-affected zone.

Franks [3] graphically shows that heat input is an important factor in determining the cooling rate, and that high heat inputs lessen the rate of cooling. However, heat input, having an upper limit beyond which the heat-affected zone is adversely affected, probably cannot adequately control the cooling rate in welds made in steels with high hardenability.

SECTION 5

THE WELDABILITY OF NICKEL-CHROMIUM-MOLYBDENUM STEELS

The principal difference in the welding behaviour of alloy and plain-carbon steels is the susceptibility of the alloy steels to cracking in the heat-affected zone during or after cooling, a consequence of their higher hardenability.

All of the alloying elements affect weldability in the proportion by which they affect hardenability. Consequently, the more highly alloyed the steel the more necessary it is to lower the carbon content or to use procedures of preheating and postheating in welding which are normally used for higher carbon steels.

A long-term investigation of cracking in welded joints of low carbon, low-alloy steels is reported in Reference 6. This includes an extensive literature search of some 92 related articles on the weldability of a great variety of low-alloy steels, in addition to the author's investigation. The latter is particularly pertinent to the writer's investigation in that it includes two nickel-chromium-molybdenum steels very close to modified HY-80 in composition, one with .20 percent carbon, one with .23 percent carbon.

Berry and Allan [6] have made some general observations for steels susceptible to heat-affected zone cracking which can be summarized as follows:

a. Martensite is essential to heat-affected zone cracking.

b. Cracks often originate in a "mixed structure" of martensite and bainite and are found at the boundaries between these structures, not necessarily the previous austenite grain boundary. Thus the crack separates a plastic and a non-plastic region.

c. The "mixed structure," showing the high susceptibility to cracking occurs in the heat-affected zone of steels in which the martensite transformation is complete (M_f) between approximately 100 C and 250 C.

d. The two nickel-chromium-molybdenum steels show a great susceptibility to heat-affected zone cracking at relatively high cooling rates.

e. The two nickel-chromium-molybdenum steels show considerable segregational effects, or inhomogeneity. Segregational effects play an important role in weld cracking.

f. Both residual stresses and externally applied stresses are responsible for heat-affected zone cracking.

g. Decomposition of austenite at room temperature was not an essential factor in weld cracking.

h. Cracks usually originate at stress raisers such as sulfide or oxide inclusions, porosity, etc.

i. Hydrogen appears to play its major role during crack propagation, after an embryo crack has formed.

j. Heat-affected zone cracking was present in one nickel-chromium-molybdenum steel even in an atmosphere of argon or CO_2 when the low temperature cooling rate was as high as 9 degrees F/sec. and upward. In partial hydrogen atmospheres the cracking increased as hydrogen partial pressure increased.

k. Vanadium proved to be a good replacement for molybdenum in alleviating temper brittleness in steel, and better mechanical properties in welds resulted.

l. The principal structural features, therefore the mechanical properties, depended primarily upon the cooling rate.

m. The low temperature cooling rate was correlated to weld cracking, a lower cooling rate decreasing the incidence of heat-affected zone cracking.

n. Moderate preheating had a marked effect on slowing the low temperature cooling rate and the heat-affected zone cracking, the safe cooling rate depending on composition, atmosphere, and welding conditions.

The authors concluded that, using a normal heat input, both the .20 percent carbon and the .23 percent carbon nickel-chromium-molybdenum steels were "virtually unweldable" without preheating. They recommended that for "universal weldability," when preheating is not practicable, that the carbon be lowered, and the alloying be increased as needed.

Rossi, in discussing the weldability of alloy steels [5], states:

> The nickel-chromium-molybdenum steels containing over 3 percent nickel and 1 percent chromium require special welding precautions. These steels are very prone to crack in the heat-affected zone even under moderate restraint. However, by regulating the preheat and interpass temperature to within the temperature range of 450-600 F, and by maintaining this temperature after welding for about 10 to 12 hours, the welds are rendered completely satisfactory.

Rossi further points out that the nickel-chromium-molybdenum steels with .12 - .20 percent carbon and less alloy content are less critical to weld, requiring only a preheating temperature of 200-300 F. For higher carbon contents (modified HY-80 has a maximum of .23 percent carbon), he recommends a higher preheating temperature for these steels.

Fuchs and Bradley [17], point out that allow steels containing, singly or in combination, .3 - .4 percent carbon, nickel up to 3.5 percent, chromium up to 1.5 percent, molybdenum up to .6 percent, and manganese up to 1.5 percent, are characterized by the fact that the production of satisfactory welds usually requires special precautions such as preheating and postheating, and the use of austenitic electrodes.

Cahill, [3], in discussing successful welding of regular HY-80 steel, states:

> While we have had little or no trouble in weld fabrication of high yield strength steel (generally about 1" maximum thickness) on our carriers at New York, it is to be remembered that it has always been our practice to preheat with electric strip heaters on all our high yield strength steel and special treatment steel work.

Cunningham, [3], comments on the welding of heavy plate modified HY-80 steel:

> It became apparent that the material with which we were working, at least in thicknesses above 1¼", was more closely akin to special treatment steel than we had at first realized. This realization pointed to the need for higher preheat. We feel that the precautions which are necessary are pushing the limits of practical shipbuilding. We feel that basically the difficulties stem from the chemistry of the HY-80 parent material, particularly in plates over 51 lbs/ft^2. Actually, the two ranges of chemistry specified result in two entirely different grades of steel, which, although similar physically, are far apart in weldability.

Mr. W. Ericksen, Head Metallurgist Westinghouse Sunnyvale, California Plant, makes the following statement concerning the welding of heavy plate modified HY-80:

> To render our modified HY-80 welds made with type 11018 electrode tough and crack-free we use a preheat of 150-200 F and stress-relief annealing at 1075 F for one hour for each one inch of plate thickness.

SECTION 6

OTHER RESEARCH ON THIS PROBLEM

The Naval Research Laboratory, naval and industrial shipyards, and industrial welding equipment companies have all participated in a search for materials, methods, and techniques which would decrease the high incidence of cracking in the welded joints of both the modified and regular HY-80 steels. Considerable attention has been given to the variables of (1) heat input, (2) preheat, (3) welding method and technique, and (4) weld metal and flux composition.

Reference 3, a compilation of reports from various activities working on this problem, discusses in detail the effect of the many variables noted above. Some of the more important findings reported by these activities are as follows:

THE BASE METAL

a. The two ranges of composition (regular HY-80 and modified HY-80) result in two entirely different grades of steel which are far apart in weldability.

b. Specification quality HY-80 steel base plate is found to be highly resistant to fracture in explosion bulge testing at test temperatures down to 0 F. This is true for both regular HY-80 and modified HY-80.

c. No cracks were observed to penetrate into the unaffected region of the plate when testing welds by explosive bulge tests.

d. The harder the heat-affected zone, the more sensitive to cracking it becomes.

WELDING ELECTRODE

a. Low moisture content of the coatings is of primary importance in controlling cracking.

b. In most cases low hydrogen type 11018 weld rod showed the most promise for arc welding of modified HY-80 steel.

WELDING METHODS AND TECHNIQUE

a. Manual arc welding method appears to be the most generally reliable method of welding modified HY-80 steel.

b. Inert gas welds generally give higher notch toughness of weld deposits above -40 F than do submerged arc welds.

PREHEATING, HEAT INPUT, AND INTERPASS TEMPERATURE

a. Both rate of arc progression and welding current had a marked effect on the cooling rate of the heat-affected zone, hence effect on grain size and mechanical properties.

b. The heat-affected zone is adversely affected by excessive heat input, generally showing more deterioration at 80,000 joules/in. than at 40,000 - 50,000.

c. Relatively high heat inputs are necessary to meet radiographic standards.

d. A heat input of approximately 40,000 joules/in. appeared to be the least detrimental to the heat-affected zone in submerged arc welding.

e. Weld deposit notch toughness is little affected by varying heat input, arc voltage, amperage, and rate of progression.

f. Preheating in the vicinity of 300 F has a pronounced effect on the low temperature cooling rate of the heat-affected zone. A relatively low preheating temperature has little effect on heat-affected zone hardness, but does decrease its susceptibility to cracking.

From a study of this comprehensive report one concludes that preheating and heat input play major roles in controlling the microstructure, thus the mechanical properties of the heat-affected zone.

SECTION 7

EXPERIMENTAL WORK

GENERAL

After consultations with personnel directly engaged in new submarine construction and after a study of the history of cracking in the welded joints of modified HY-80 steel, the decision was made to restrict the study principally to the heat-affected zone of the base plate, and to compare the effect of welding without preheating and of preheating to 150 F on this zone.

The purpose of this work was to determine the primary cause of heat-affected zone cracks and to examine the contributory factors to cracking in the absence of externally applied stress.

Welded specimens were furnished by Mare Island Naval Shipyard, fabricated in the welding engineer's test laboratory. Figure 7-1 depicts these specimens and the manner in which they were fabricated. Welding procedure was as stipulated in BuShips Notice 9110, except that some specimens were made with no preheating. The base plate from which specimens were taken was 1 13/16" modified HY-80, a fairly large section (3' X 4') in order to realize a realistic heat dissipation, simulating that in pressure hull welding. The base plate was unrestrained during welding. The composition of the plate by chemical analysis in weight percent was found to be as follows:

C	Mn	P	S	Si	Cr	Ni	Mo	Va
.16	.29	.008	.01	.22	1.40	2.91	.46	.01

Manual arc welding was used with type 11018 electrode, which had been baked to eliminate hydrogen. Heat input in welding was maintained at 50,000 - 60,000 joules/inch.

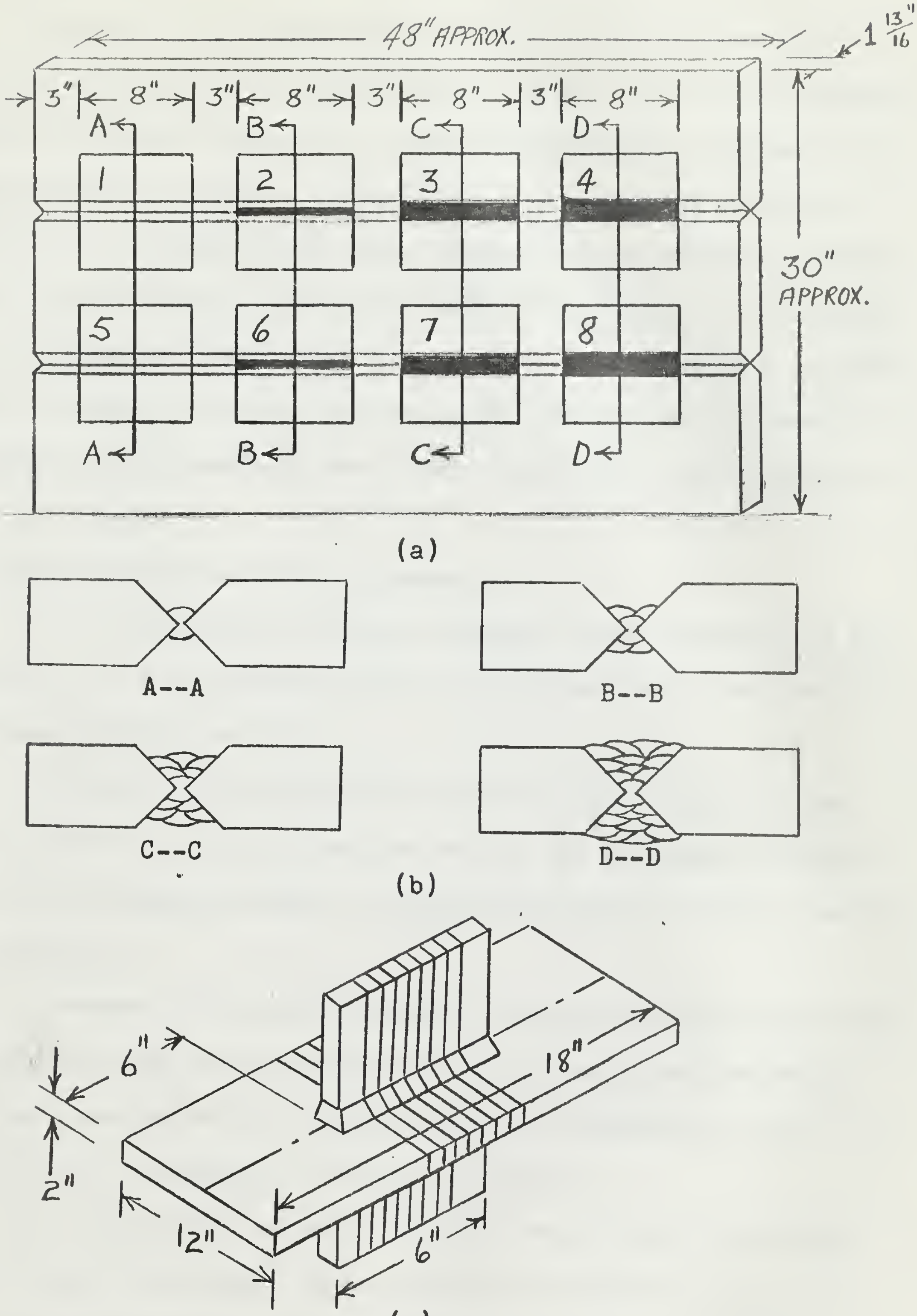


Figure 7-1. Details of welded specimens. (a): Butt welded specimens 1 through 8. (b): Cross-section of butt welds. (c) "T" joint specimens.

NOTCH TOUGHNESS

These tests were carried out to compare the relative ease of fracture under impact loading of the heat-affected zone, the weld metal, and the unaffected base metal. Standard Charpy V-notch specimens were used throughout the test. In evaluating notch toughness in the heat-affected zone, specimens were taken perpendicular to this zone with the notch located exactly in the coarse grain area (Figure 7-2). Specimens from the base metal were taken both parallel and transverse to the direction of rolling, all specimens being taken from the same horizontal plane. Weld deposit specimens were taken both longitudinally and transverse to the line of weld.

These tests revealed the following:

a. A pronounced loss in notch toughness in the heat-affected zone (compared to the as-received plate) in all welded joints, particularly those made without preheating.

b. Higher notch toughness for as-received base plate specimens taken parallel to rolling direction than for the transverse specimens. (This is particularly evident in comparing specimens taken near the rolling surface.)

c. Somewhat higher notch toughness in weld deposit specimens taken transverse to line of weld compared with the longitudinal specimens.

Key quantitative data from these tests is presented in Figure 7-3. Complete data is presented in Table 1, Appendix I.

It is significant that notch toughness values found in the heat-affected zone, particularly those made without preheating, are only a fraction of the toughness found in the unaffected plate. It will be shown later that the heat-affected zone is by far the hardest zone, and possesses very poor ductility as indicated by tensile tests.

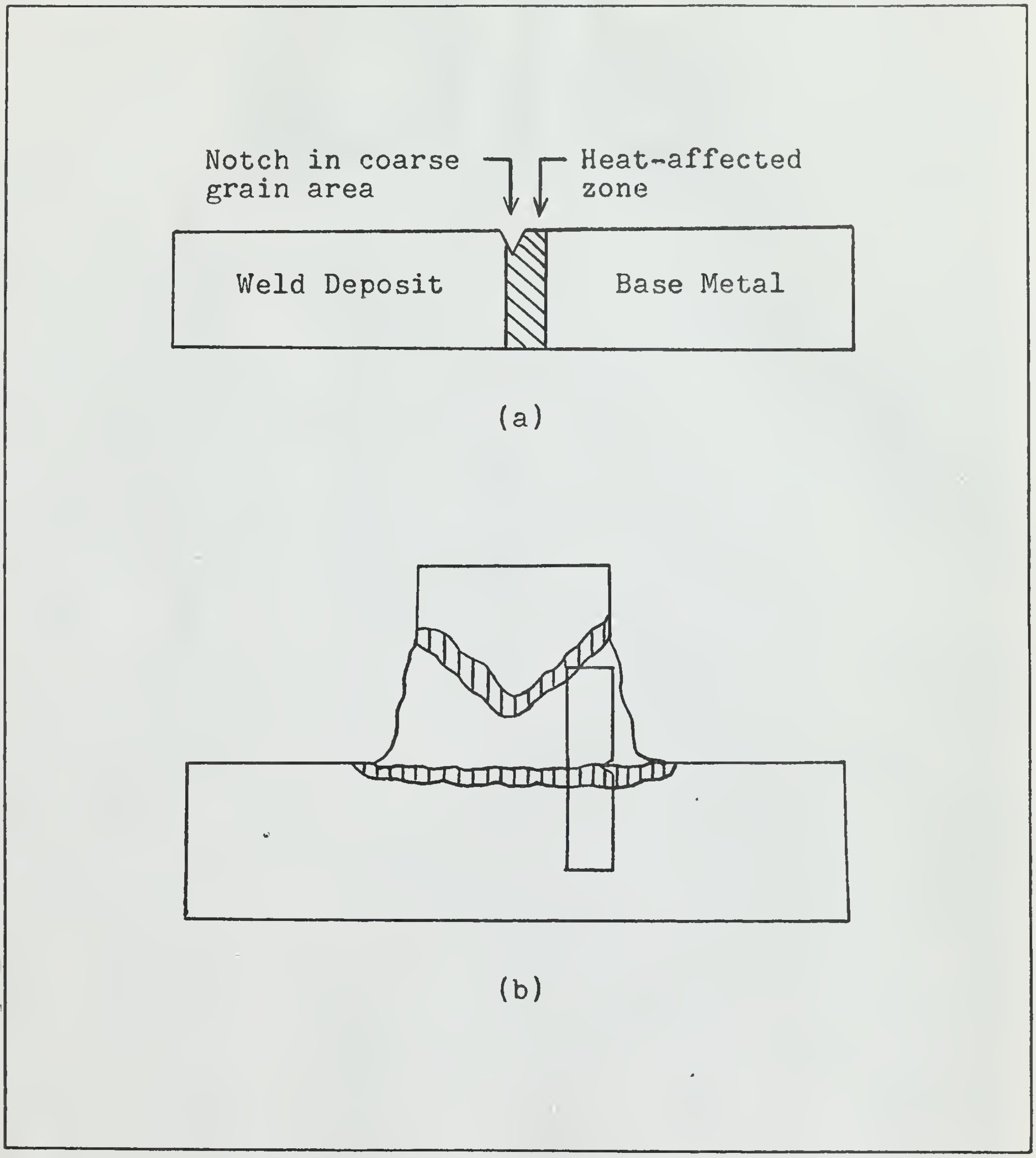


Figure 7-2. Impact test specimen for measuring the notch toughness of the heat-affected zone. (a): Standard Charpy V-Notch test specimen. (b): Method for taking specimen from welded "T" joint.

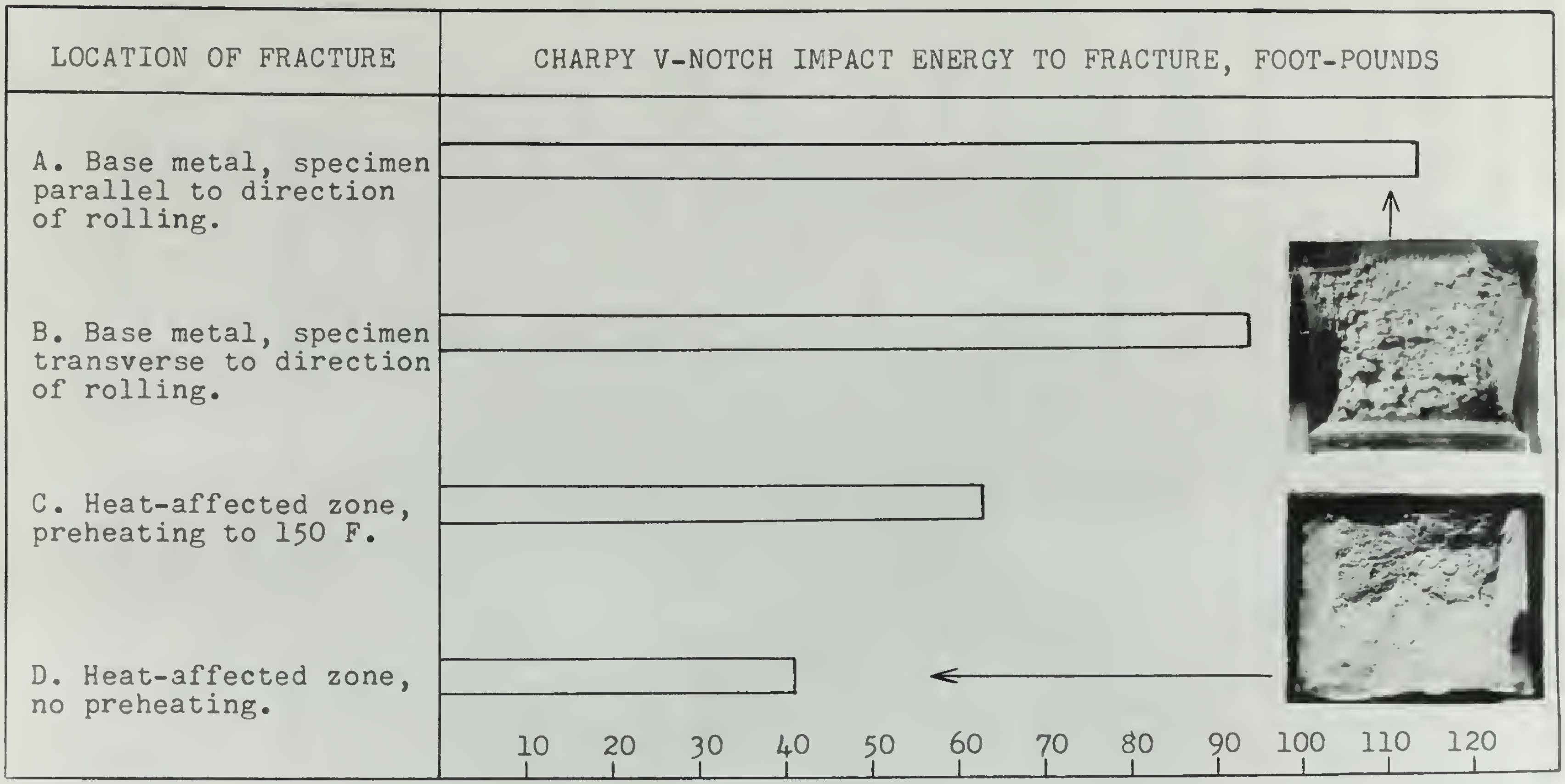


Figure 7-3. Comparison of notch toughness of the heat-affected zone with that of the as-received modified HY-80 base plate. Measurements shown represent the average of three specimens fractured at 72 F. Upper photo shows fracture surface of unaffected metal, showing fibrous structure and considerable lateral contraction. Lower photo, fracture in the heat-affected zone, shows granular appearance, negligible lateral contraction.

The heat-affected zone of the welds made with a preheating temperature of 150 F showed greater notch toughness than those made without preheating. However, the fracture energy of the former was still only approximately 60% of that in the unaffected base plate.

Results of impact testing have been found to correlate closely with actual ship-board service failures [10, 20].

HARDNESS MEASUREMENTS

In steel, high hardness in a welded joint generally connotes embrittlement, thus susceptibility to cracking on deformation. Therefore, extensive hardness readings were taken in this investigation.

Hardness measurements were made using a Tukon Hardness Tester, with a 1 kg. load, on specimens polished and etched as if for micro-photography. Because of the small area involved and the varying hardness within the zone, micro-hardness readings proved to be the most satisfactory means of accurately measuring the hardnesses within the heat-affected zone. Since the heat-affected zone cracks often appear as micro-cracks (within a small region) in a hardened area, it follows that micro-hardness measurement is the best means to correlate hardness with micro-cracking.

The average Knoop Hardness Number in the heat-affected zone adjacent to the upper ½" of weld deposit was 418, comparable to Rockwell "C" 41 hardness, in welds made without preheating. The upper ¼" was even harder with an average Knoop Hardness Number of 438, comparable to Rockwell "C" 43. The maximum reading was Knoop Hardness Number 508. The heat-affected zone was somewhat softer near the root passes. Too, there were occasional relatively soft spots due to some tempering effect of the multi-pass welding.

In the welds preheated to 150 F, the average hardness in the heat-affected zone was, surprisingly enough, almost identical to that in the non-preheated welds. Preheating at 150 F apparently is effective in furnishing some stress-relief, for cracking is reportedly decreased with this preheating temperature, although little difference in hardness is noted.

Weld deposit average hardness in non-preheated welds was approximately Knoop Hardness Number 325, with higher hardness readings in the upper weld passes. Weld deposit immediately adjacent to the upper heat-affected zone was sometimes as hard as Knoop Hardness Number 370, thus explaining the relative ease of crack propagation into this area. Weld deposit hardness in the preheated welds was only slightly less.

The average hardness in the base metal was Knoop Hardness Number 267, the hardness being fairly uniform in the unaffected plate.

Figure 7-4 graphically illustrates typical hardness readings across a butt welded joint, showing hardness readings for specimens welded with and without preheating, and for one weld which had been tempered at 1150 F after welding. It is particularly significant to note both the high hardness and width of the hardened zone of the welds made with and without preheating. A generally accepted relationship between hardness, carbon content, and percent of martensitic structure present is shown in Figure 7-5. A wholly martensitic structure in this steel (.16 carbon) would show a hardness of Rockwell "C" 42. Microscopic examination of the heat-affected zone adjacent to the upper weld passes reveals, indeed, that this structure is very high in martensite. Figure 7-6 shows a typical microstructure in the heat-affected zone of a modified HY-80 weld with no preheating.

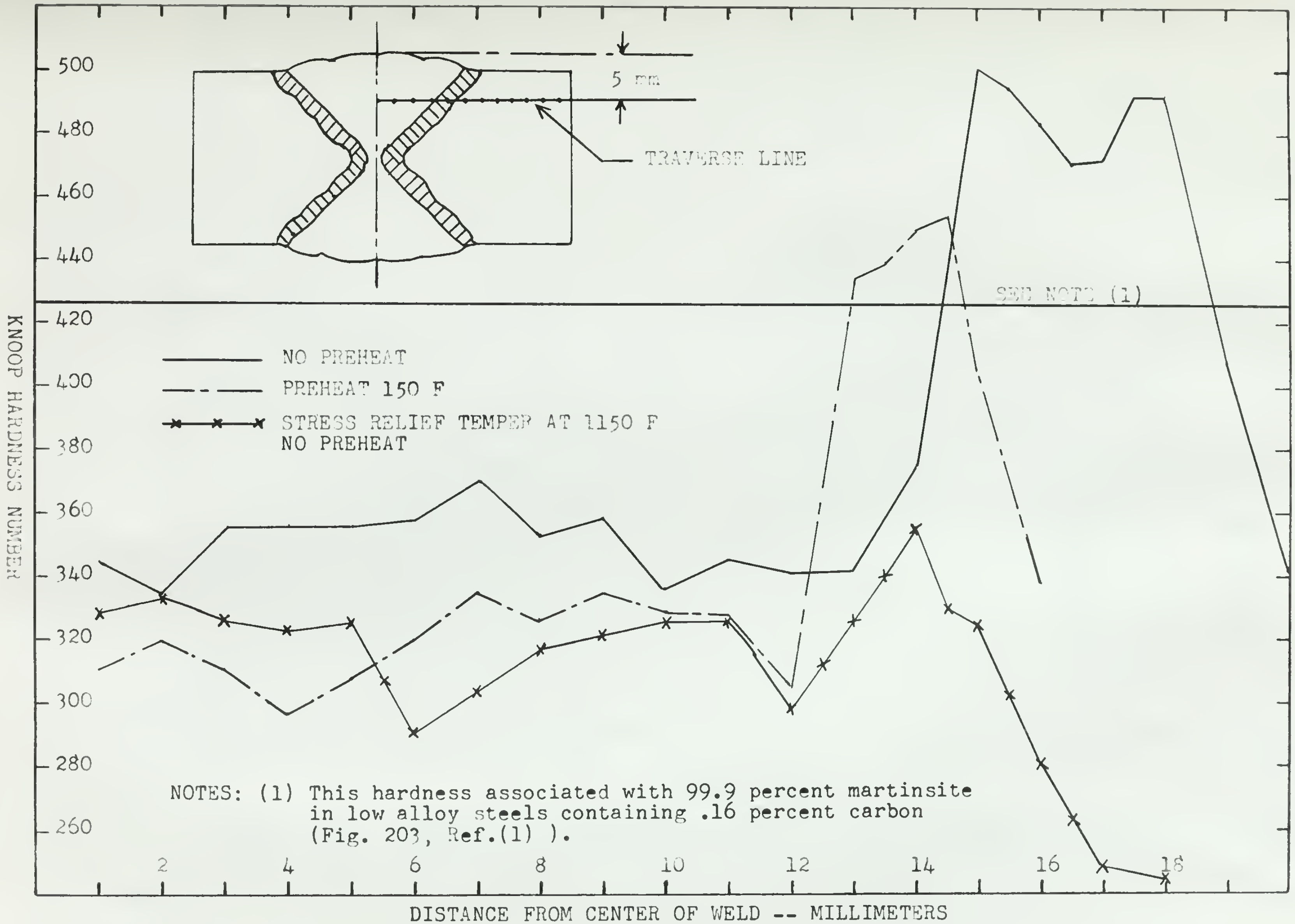


Figure 7-4. Hardness measurements across the welded joint of modified HY-80 steel, type 11018 electrode.

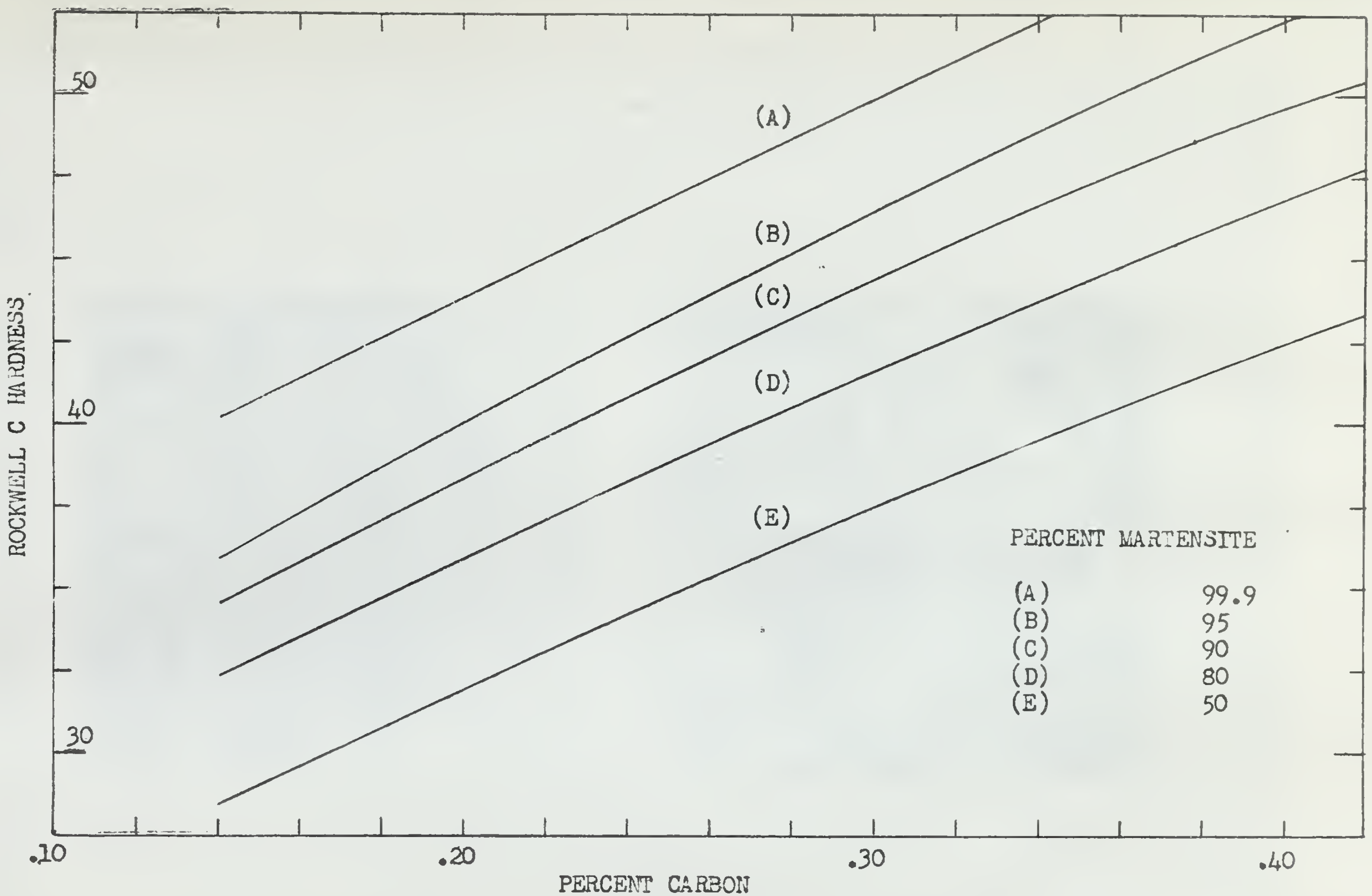


Figure 7-5. Average relationships between carbon content, hardness, and percentage of martensite for several low alloy steels.

Figure 7-6. Typical microstructure in the heat-affected zone of modified HY-80 steel, welded without preheating. (250X)

Figure 7-7 illustrates the location of the weld deposit, the heat-affected zone and the unaffected base plate in relation to high hardness readings.

Tabulations of hardness readings taken for the various welds are presented in Tables 2A, 2B, and 3, Appendix I.

Micro-hardness tests were also conducted in "T" joint welds made by (1) the MIG welding method, using type B88 electrode, and (2) manual arc method, using type 9018 electrode. These welds were made under conditions specified by the Bureau of Ships using 50,000 - 60,000 joules/inch of heat input. These readings are presented in Table 3, Appendix I. The following observations are made from these tests:

a. The average heat-affected zone hardnesses for both welds compare closely with those using type 11018 electrode when using the same preheating temperature.

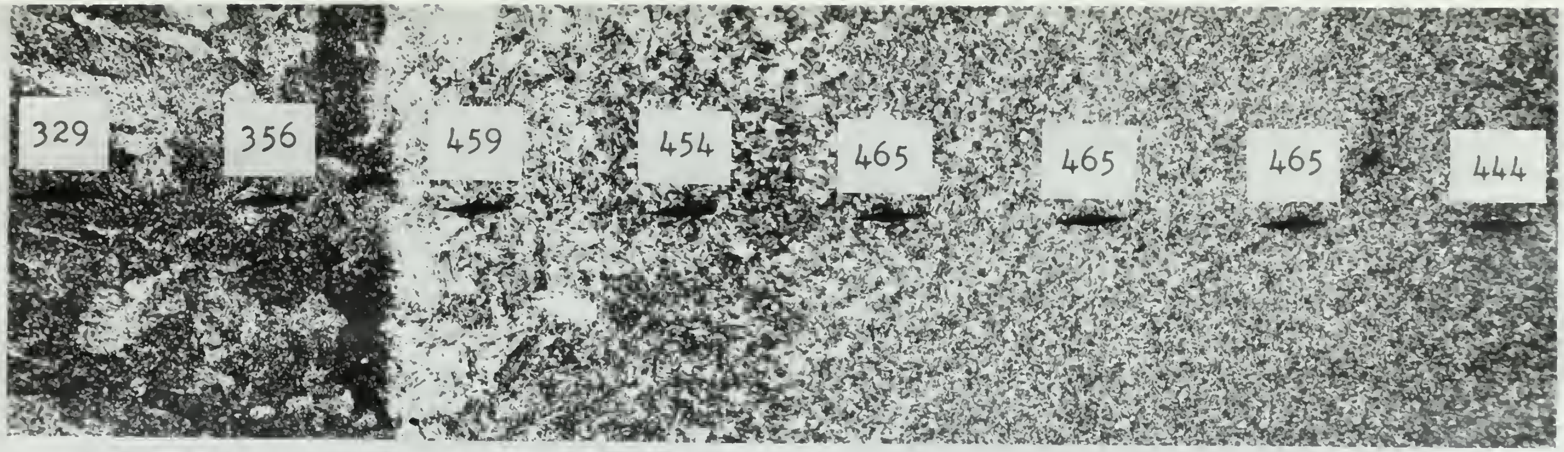
b. The average hardness of the type B88 weld deposit is considerably greater than that of type 9018.

c. The hardness readings in type B88 weld deposit were the more uniform of the two; also considerably more uniform than type 11018.

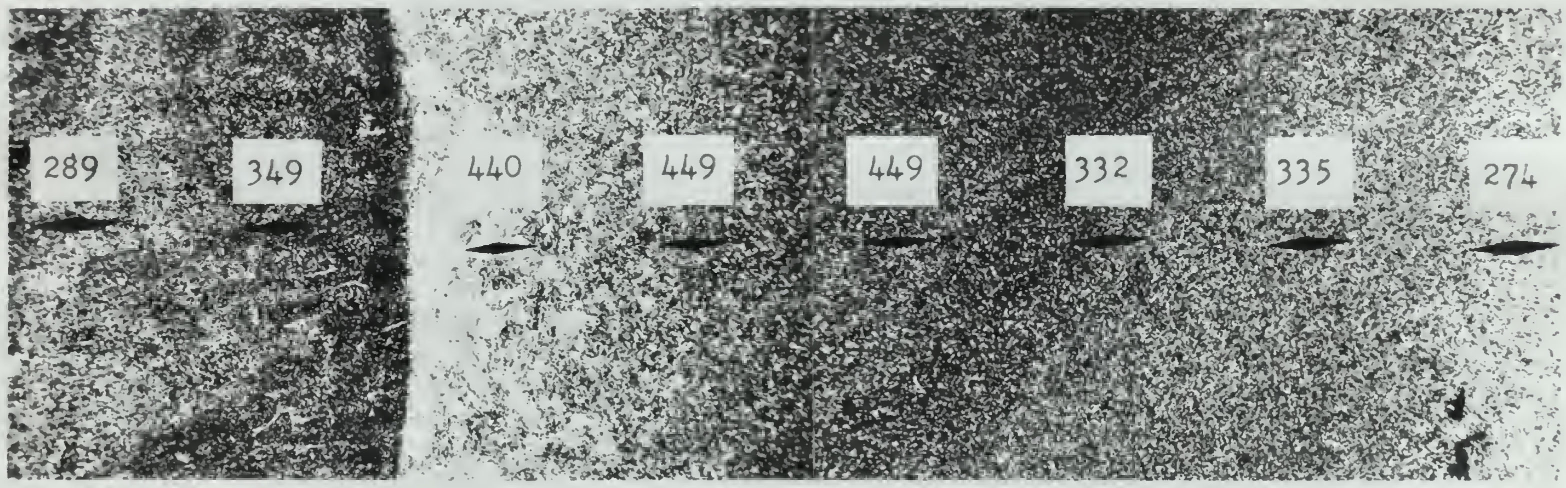
MICROSCOPIC EXAMINATION

Several micro-photographs of the sites of heat-affected zone cracking are shown in Figure 7-8. Generally, these occur predominantly in (1) the areas of grain growth, (2) martensitic-bainitic mixed structure, apparently high in martensite content, and (3) in the boundaries separating these structures, not necessarily the previously austenitic grain boundary. These boundaries have been called lath boundaries by one investigator [6].

Examination of the heat-affected zone for a grain boundary precipitate, mentioned in some reports, revealed two network structures. One

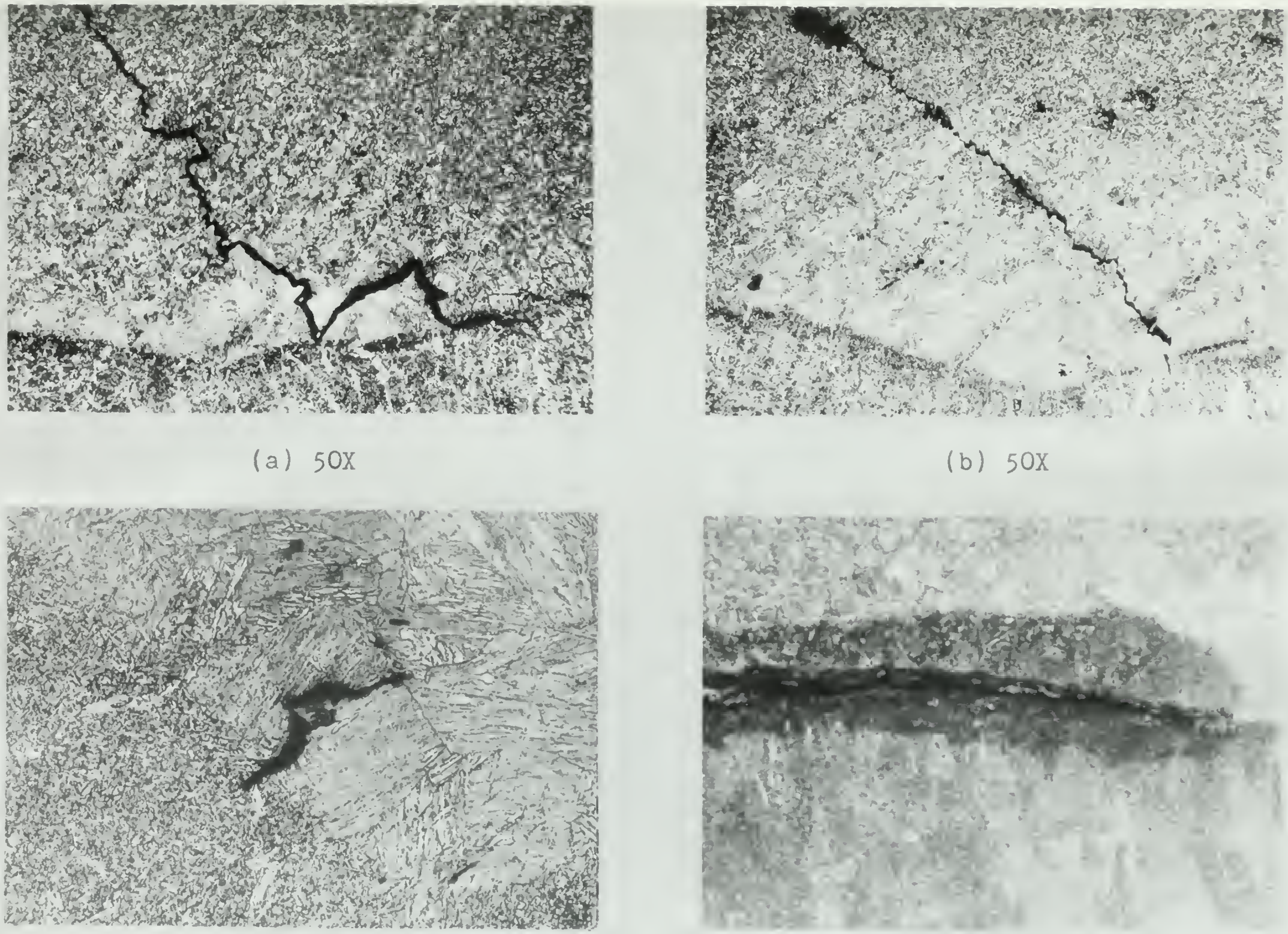


(a) Welded without preheating



(b) Preheated to 150 F

Figure 7-7. Heat-affected zone and weld deposit, showing site and values of hardness measurements. Numbers shown are Knoop hardness numbers. Readings are .5 mm apart. (50X)



(a) 50X

(b) 50X

(c) 250X

(d) 50X

Figure 7-8. Typical cracks occurring in heat-affected zone and vicinity. Weld deposit is lower portion of each of above photos. Note in (a), (b), and (c) that cracks terminate on entering weld deposit (typical, but not always the case). (d) shows fusion zone crack.

structure appears to be a network of globular carbides surrounding an area of carbides plus tempered martensite. The second structure appears to be a network of untempered martensite surrounding tempered martensite. No cracks were observed to propagate through such areas. A few cracks were observed to enter such areas and terminate shortly thereafter; however, this may not always be the case. In the specimens examined, these network structures occurred only sparingly.

Weld deposit and heat-affected zone cracking occurred spontaneously some three weeks after completion of welding in all of the non-preheated welds which were incomplete; i.e., where several weld passes had been made, but the final passes were missing. Figure 7-9, showing a longitudinal slice through one of these welds, illustrates the extent of spontaneous cracking. This is considered particularly revealing, inasmuch as the welds were made in unrestrained plates, and no restraint was ever applied to these joints subsequent to welding. The spontaneous cracking indicates that the stress distribution has changed over the welded joint, and is presumably due to a phase change or a redistribution of point defects after completion of welding.

The weld deposit is characterized throughout by the columnar structure characteristic of a very rapidly quenched steel casting. Alternate high and low carbide areas give a pronounced segregated appearance. Weld deposit cracks often appear in the boundaries separating these segregated areas. This is depicted in Enclosure 51, to Reference 13. Many of the weld deposit cracks appeared to originate in the vicinity of the fusion zone, often propagating a short distance through the weld deposit and terminating prior to reaching the outer surface. Figure 7-9 is typical of this phenomenon.

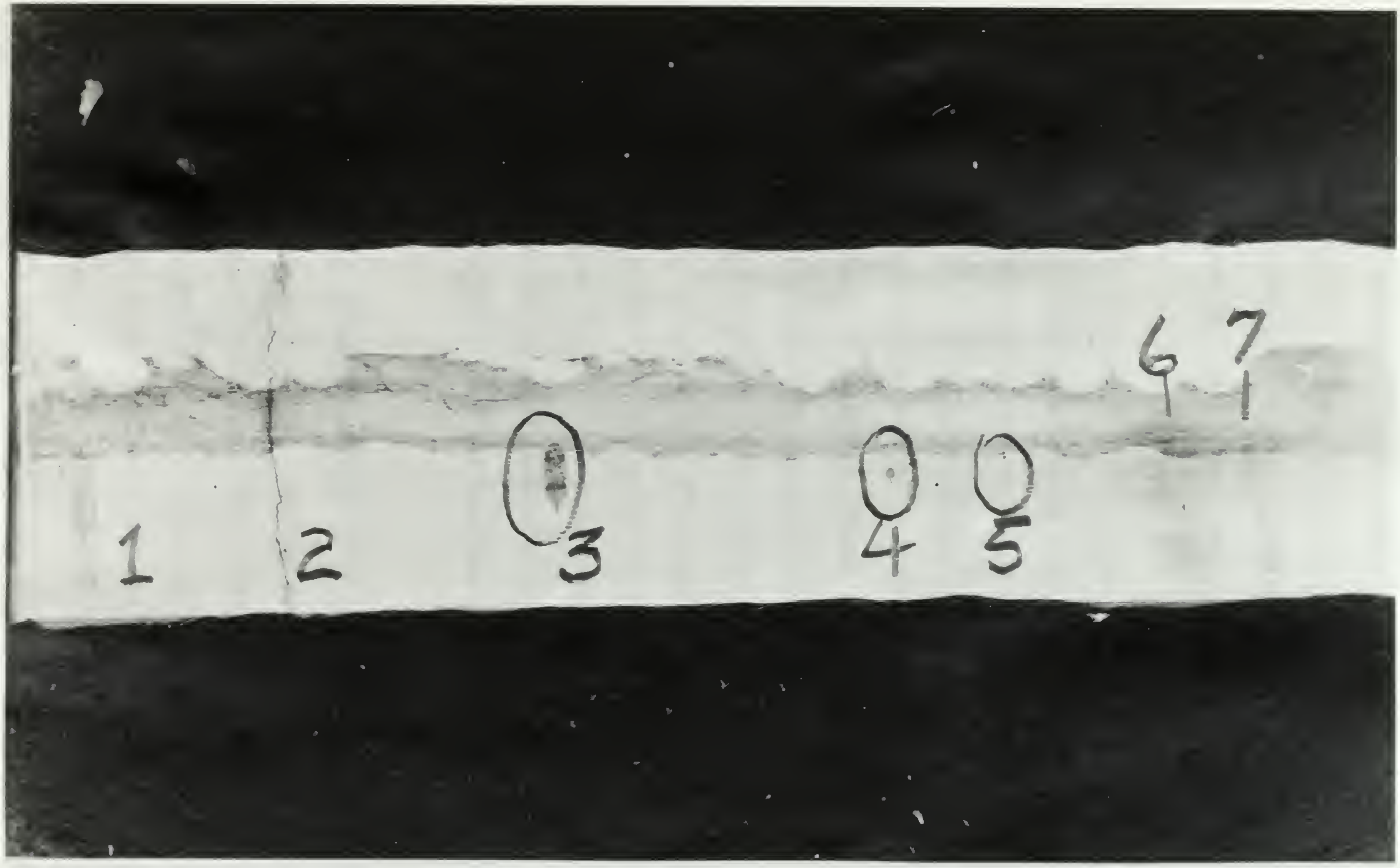


Figure 7-9. Spontaneous cracking in a welded joint of modified HY-80 steel, type 11018 electrode. Plate was unrestrained, welded without preheating.

Evidently many weld deposit cracks occur at a fairly high temperature. In several instances, trepan sections of weld cracks showed dark blue-peacock coloration typical of high temperature oxidation. Sometimes this coloration is present in only a portion of the crack--the remainder of the surface appearing relatively clean and bright, characteristic of a cold crack.

The base metal appears relatively free of non-metallic inclusions. However, inclusions resembling manganese sulfide and iron ixide are present in small amounts, and the former occasionally occurs in a chain-like growth. In view of the high residual stresses in the heat-affected zone, no great amount of inclusions is needed to facilitate crack initiation. Since the steel is relatively clean, no further emphasis need be put on minimizing inclusions to prevent cracking.

The spontaneous cracking in the welded joint of non-preheated specimens emphasizes the development of extremely high residual stresses, and leads one to suspect the possibility of retained austenite, transforming in time to martensite with a volumetric change and attendant additional stresses. However, no retained austenite was detected visually in the welded joints or in fully hardened specimens of base plate and weld deposit.

PROPERTIES OF FULLY HARDENED MODIFIED HY-80 AND THE EFFECTS OF TEMPERING

This experimentation was for the purpose of evaluating (1) the degree of embrittlement attainable in modified HY-80, and (2) the effects of tempering on notch toughness, ductility, and hardness. Since a diversity of opinion exists as to the degree of brittleness attainable in a low carbon steel, this study should be pertinent to the problem. If results indicated that fully hardened modified HY-80 is very low in ductility and in notch toughness, and if the heat-affected zone microstructure and

properties approximated those of the fully hardened steel, then a high incidence of cracking in the heat-affected zone could be expected.

Standard Charpy V-notch impact specimens and tensile specimens were machined as specified in Reference 2. These were all taken transverse to the direction of rolling. The austenitizing treatment was carried out at 1800 F in a controlled atmosphere furnace to prevent decarburization. All specimens were oil quenched, then tempered at various temperatures from 400 F to 1150 F. The notches in the impact specimens were machined after quenching and tempering to prevent possible crack initiation in the heat treatment. Figure 7-10 graphically summarizes the results of these tests.

It is obvious from these results that fully hardened modified HY-80 has very low notch toughness fracture energy, practically nil-ductility, and relatively high brittleness. One further observes that tempering for two hours at the temperatures indicated in Figure 7-10 has little effect on these values and on hardness until the tempering temperature is raised to approximately 800 F, where toughness and ductility increase rapidly as the material softens. Figures 7-11 and 7-12 show the fracture surfaces of fully hardened base plate, comparing them with fractures in as-received plate. Note that the fractures in fully hardened specimens show virtually no lateral contraction.

EFFECTS OF ANNEALING AS-RECEIVED BASE PLATE

Charpy V-notch impact test specimens were used to determine, if any, the extent of embrittlement of previously quenched and tempered modified HY-80 base plate upon annealing in the range of 400 to 1000 F. Test specimens were taken both parallel to and transverse to direction of rolling. The specimens were annealed for one hour at various temperatures,

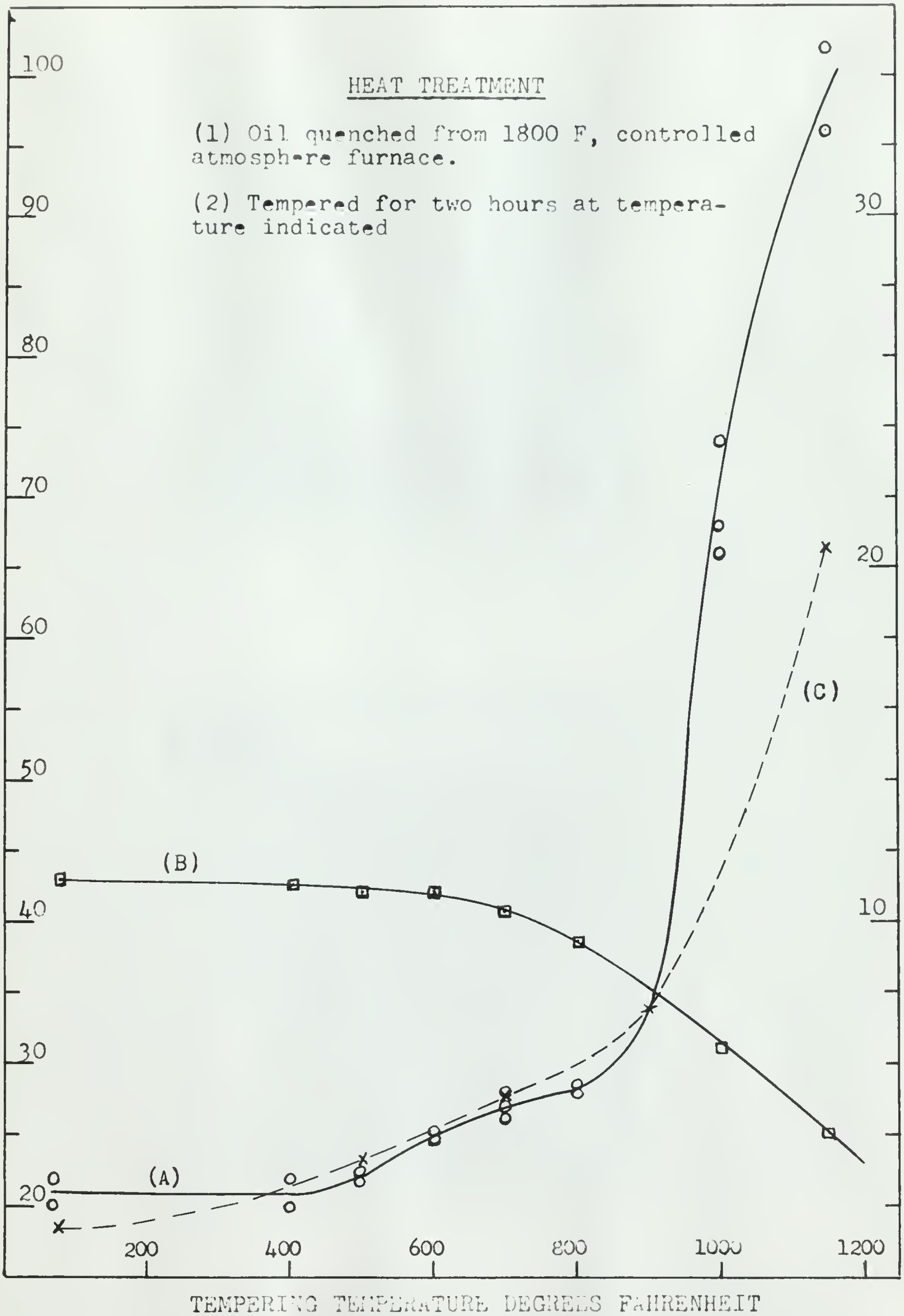


Figure 7-10. Effects of tempering on the mechanical properties of fully hardened modified HY-80 steel. (A): impact energy, (B): hardness, (C): elongation.

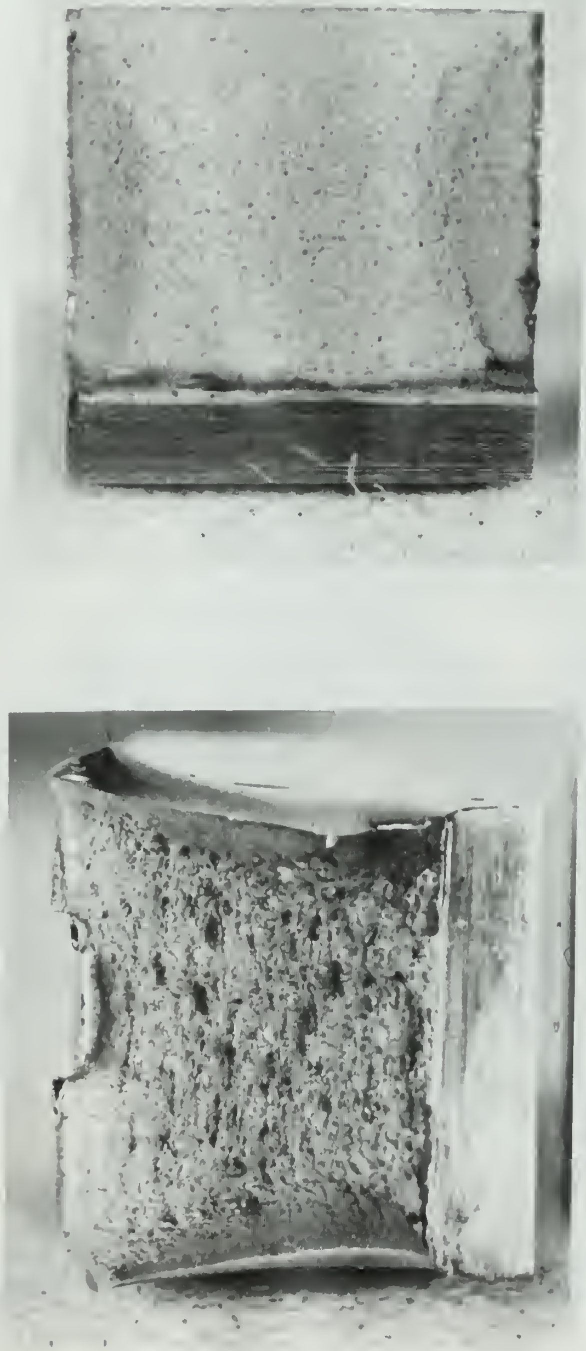


Figure 7-11. Impact fracture surfaces of modified HY-80 base plate. Upper: Fully hardened, showing granular appearance and negligible lateral contraction at break. Lower: As-received base plate, showing fibrous appearance and considerable lateral contraction (6X).

Figure 7-12. Fractured tensile specimens of modified HY-80 steel. Upper: as-received base plate, showing "necking down", ductile type fracture. Lower: fully hardened plate, negligible "necking down", brittle type fracture.

then cooled to room temperature for testing. The results of this experiment are tabulated in Table 4, Appendix I.

Little or no loss in impact energy was noted, except for specimens cut perpendicular to the direction of rolling and tempered at 400 F. Later, in attempting to re-evaluate this phenomenon, similar tests revealed impact energies of 105, and 106 ft-lbs, respectively, for two test specimens tempered at 400 F. In attempting to determine the reason for the considerable difference between the first and last tests, it was then learned that the transverse specimens for the original test had been taken from the portion of the base plate near the rolling surface, the latter specimens from the central portion of the plate. Time did not permit further tests, but these tests indicate inhomogeneity across the thickness of heavy plate modified HY-80.

TENSILE TESTS TO EVALUATE THE DUCTILITY OF THE HEAT-AFFECTED ZONE

Figure 7-13 shows two tensile specimens, both for evaluating the ductility of the heat-affected zone. One is taken transverse to the weld joint, the other with the heat-affected zone oriented longitudinally in the test specimen.[1] A small heat-affected zone crack is visible, occurring in this case after an elongation of 4%. Table 5 tabulates data for these measurements.

Thus, in three non-preheated specimens, the average elongation prior to visible cracking in the heat-affected zone was 3.33%. The preheated specimens showed somewhat more ductility, an average of 8.25% elongation. However, Figure 7-14 shows preheated specimen number 5, in which a heat-affected zone crack appeared after only .5% elongation. It is particularly

[1] The writer concurs with Parker [10] that the latter is the better specimen for evaluating the ductility of the heat-affected zone.

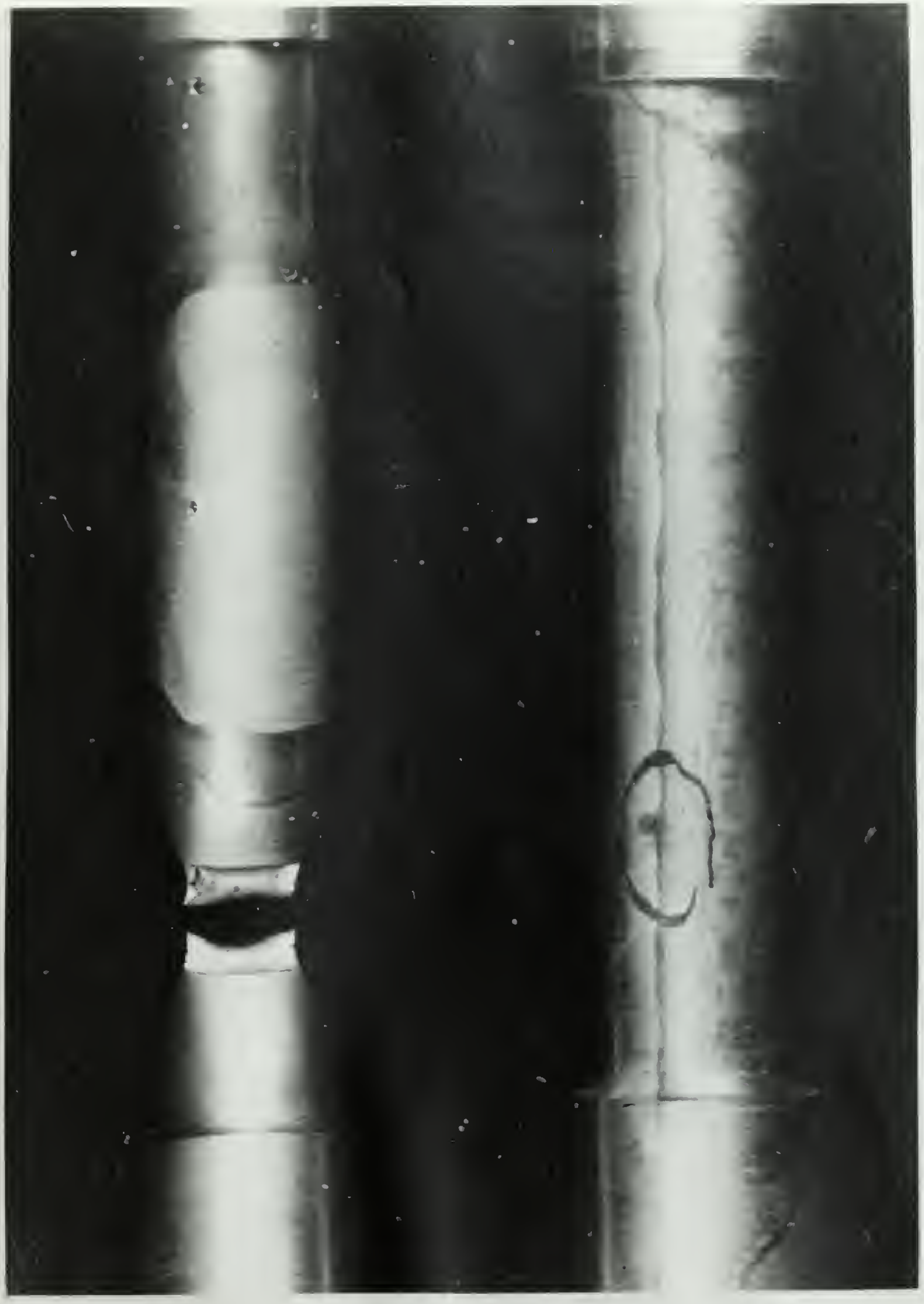


Figure 7-13. Tensile specimens for measuring the ductility of the heat-affected zone of a welded joint. Modified HY-80 steel, type 11018 electrode. Left: heat-affected zone oriented transversely, fracture occurs in base plate. Right: heat-affected zone longitudinal, crack occurs in heat-affected zone.

Figure 7-14. Heat-affected zone crack in tensile specimen preheated to 150F. Modified HY-80 steel, type 11018 electrode (8X).

significant that some specimens, whether non-preheated or preheated to 150 F do crack in the heat-affected zone after negligible plastic deformation.

STRAIN-AGING TESTS

These tests were conducted to examine the tendency for modified HY-80 steel to become embrittled due to cold working during forming and subsequent heating. Since Reference 8 indicated some embrittlement following the strain-aging of normalized HY-80 steel at lower-than-ambient temperatures, these tests were conducted at 32 F. Flat bars were machined from the base plate, transverse to direction of rolling, then reduced 5% by a further rolling along the length of the bar. Impact specimens were then taken lengthwise from the bar. The finished Charpy V-notch specimens were then heated for one hour at various temperatures, then air cooled. Results of the impact tests are given in Table 6, Appendix I.

Thus some embrittlement is indicated in base metal which has been deformed plastically then later heated in welding or burning operations. Steels susceptible to this embrittlement are embrittled whether the plastic deformation occurs before or during the heating [1].

RETAINED AUSTENITE

Incidents of delayed cracking suggest the possibility of retained austenite which transforms under stress to martensite, a volumetric change. This is not uncommon in alloy steels.

A great portion of the time devoted to this research was devoted to x-ray diffraction tests of both base metal and weld metal in an attempt to detect any austenite present.

The procedure was first to calculate the angles and relative intensities to be expected from face-centered cubic austenite, using a copper x-ray target. The Bragg angle to be expected for each set of reflecting planes was verified approximately by running a diffractometer check on a purely austenitic sample of stainless steel.

A pulse height analyzer, with a controllable "window", providing wave length discrimination, was used in conjunction with the diffractometer. By the proper setting of the base of the window, the iron fluorescence was practically eliminated, permitting only the copper K alpha signal to appear on the diffractometer recorder.

Considerable experimentation was necessary in order to find the optimum settings, with the variables of proportional counter voltage, window, and base settings. Optimum operating conditions were determined by varying proportional counter voltage from 1550 to approximately 1800 volts, and for each of these settings varying the window and base settings to provide the best signal to background ratio. The base setting is the most critical since it is the lower edge of the window, and effectively excludes all radiation of a lower energy than that corresponding to the base setting. The window permits either a narrow or wide band of energies (above that of its base setting) depending on its setting.

Using the proportional counter and pulse height analyzer with the x-ray diffractometer, the heat-affected zone, the weld deposit, and specimens of fully hardened base plate and weld deposit were tested for the presence of austenite. No definite trace of austenite was found.

Reverting to the conventional geiger counter, diffractometer tests were repeated with the same negative results.

Because of the possibility that retained austenite existed in newly welded specimens, but transformed within a few days, one specimen was

examined by x-ray diffraction within a few hours after it had been welded. The same negative results were obtained.

It is concluded that retained austenite exists only in very small quantities--too little for detection by the pulse height analyzer technique.

It is apparent that further investigation of the possible existence of retained austenite in the heat-affected zone would have to be carried out using a crystal monochromated x-ray beam, illuminating only the heat-affected zone.

SECTION 8

DISCUSSION AND CONCLUSIONS

DISCUSSION

The literature on welding of nickel-chromium-molybdenum steels leads one to expect heat-affected zone cracking in these steels if an inadequate preheating temperature is used to control the cooling rate of the welded joint. Early investigators of this problem showed that the harder the heat-affected zone, the more sensitive it is to cracking [3]. Berry and Allan [6] correlated the low temperature cooling rate with the high incidence of heat-affected zone cracking in two nickel-chromium-molybdenum steels similar in composition to modified HY-80. They showed that high cooling rates in the vicinity of 550 - 600 F increased the incidence of cracking. Rossi [5] and Fuchs [17] both refer to nickel-chromium-molybdenum steels in general as "critical to weld" and state that they usually require both preheating and stress-relief annealing for satisfactory welded joints.

Welding engineers from numerous new submarine construction activities have related their experiences with welding modified HY-80, emphasizing that it is very difficult to weld [3].

CONCLUSIONS

This investigation was undertaken to analyze some of the metallurgical features which could be expected to lead to these difficulties in welding. Impact, tensile, and hardness measurements showed poorer properties in the heat-affected zone than in the unaffected base plate, both in specimens made with a preheating temperature of 150 F and in those made without preheating. Tensile tests demonstrated that cracks occur in these hardened zones after negligible elongation. Microscopic

examination and hardness measurements revealed that considerable untempered martensite exists in the heat-affected zone of these welded joints. Spontaneous cracking occurred in welded joints made under good shop conditions without preheating.

Assuming proper welding technique, proper control of hydrogen, and an average heat input, the primary metallurgical factors responsible for heat-affected zone cracking appear to be:

a. The high hardenability of the modified HY-80 base plate, facilitating the formation of a broad martensitic zone adjacent to the weld deposit.

b. The rapid cooling rate, especially the low temperature cooling rate, provided by the massive structure when joints are welded with insufficient preheating temperature.

c. The high restraint stresses.

The low temperature cooling rate appears to be of great importance in this problem. If M_s is high enough, and M_f is low enough, it is reasonable to predict that preheating would greatly affect the low temperature transformation products at a given cooling rate. Since, in this steel, the martensitic transformation on cooling commences at approximately 680 F, M_f is presumably approximately 200 F. A preheating temperature in this range would have a considerable effect in controlling the microstructural constituents. In addition to the effects on the microstructure, preheating in this range should be helpful in minimizing the residual stresses in the heat-affected zone. This prediction is supported in this investigation by:

a. The improvement in notch toughness and ductility as measured by tensile tests of specimens welded with a preheating temperature of 150 F, comparing them to those welded without preheating.

b. The presence of spontaneous cracking only in the welded joints made without preheating.

Although the writer finds that a preheating temperature of 150 F is helpful in decreasing the incidence of cracking in the heat-affected zone, it apparently is not sufficient to provide a welded joint free of micro-cracks. Brittle cracking in the heat-affected zone (Figure 7-14) after only .5% elongation, microhardness readings, and notch toughness tests of specimens welded with a preheating temperature of 150 F support this conclusion.

Stress-relief annealing would probably reduce the heat-affected zone cracking; however, investigation of the effects of tempering fully hardened plate indicate that a temperature in excess of 800 F is necessary to effect a considerable improvement in the mechanical properties.

Martensite, in the heat-affected zone, when adequately tempered will furnish optimum notch toughness and strength. However, the multi-pass welding operation apparently is not adequately tempering the martensitic zone in these welded joints. Therefore, some means must be employed to either decrease the formation of martensite or temper it once it has formed.

Future difficulties will probably be encountered in welding this steel under non-ideal conditions. For example, repairs alongside tenders or on foreign station, where facilities for preheating and highly skilled welders are frequently non-existent, will be extremely difficult. In fact, supervisory personnel in new submarine construction state that satisfactory single pass welds in modified HY-80 in areas not amenable to preheating are extremely difficult to make when using skilled welders.

Considering the present limitations on imposing a moderate preheating temperature, and anticipating the probable difficulties that lie

ahead once the vessel has left the yard, a change in composition of modified HY-80 base plate may be the only way in which intolerable welding difficulties can be eliminated.

In summary, the writer concludes that a moderate preheating temperature, probably greater than 150 F, is necessary for the satisfactory welding of modified HY-80 steel. In addition, when practicable, the welded joints should be stress-relieved.

BIBLIOGRAPHY

1. L. Aitchison and W. I. Pumphrey, Engineering Steels, MacDonald & Evans, LTD, London, 1953.

2. G. W. Zuspan, Basic Metallurgy, Vol. II, ASM 1957.

3. HY-80 Steel Fabrication in Submarine Construction, Report of BuShips Conference of July 1958.

4. D. K. Bullens, Steel and Its Heat Treatment, Vol. 3, John Wiley & Sons, 1949.

5. B. E. Rossi, Welding Engineering, McGraw Hill, 1954.

6. J. T. Berry, R. C. Allan, A Study of Cracking in Low-Alloy Steel Welded Joints, Welding Journal, March 1960.

7. H. Kihara and K. Masubuchi, Effect of Residual Stress on Brittle Fracture, Welding Journal, April 1959.

8. A. I. Rubin, J. H. Gross, R. D. Stout, Effect of Heat Treatment and Fabrication on Heavy-Section Pressure Vessel Steels, Welding Journal, April 1959.

9. Welding of HY-80 - General Rules for, BuShips Notice 9110, Ser 637-735 of July 1958.

10. E. R. Parker, Brittle Behavior of Engineering Structures, Wiley, 1957.

11. E. P. DeGarmo, Preheat vs Low and High-Temperature Stress-Relief Treatments, Welding Journal, May 1952.

12. B. R. Queneau, G. A. Ratz, Embrittlement of Metals, ASM, 1959.

13. H. G. MacKerrow, Mare Island Naval Shipyard Industrial Laboratory Report No. 548-59.

14. P. P. Puzak, A. J. Babecki, W. S. Pellini, Correlation of Notch Toughness with Service Failures, Welding Journal, September 1958.

15. F. J. Feely, Jr., D. Hrtko, S. R. Kleppe, M. S. Northrup, Report on Brittle Fracture Studies, Welding Journal Supplement, February 1954.

16. Handbook Committee of the American Welding Society, The Welding Handbook, 1942.

17. E. Fuchs and H. Bradley, Welding of Ferrous Metals, Vol. 2, Butterworth's Scientific Publications, 1952.

18. M.I.T. Technical Report No. 1, Contract N5-ori-78, ONR Desig. No. NR 031-256.

19. I. S. Brammar, R. W. K. Honeycombe, R. G. Ward, A New Type of Grain-Boundary Precipitate Observed in Cast Steel, Acta Metallurgica, February 1958.

APPENDIX 1

TABLE 1

NOTCH TOUGHNESS MEASUREMENTS IN MODIFIED HY-80 STEEL AND TYPE 11018 ELECTRODE

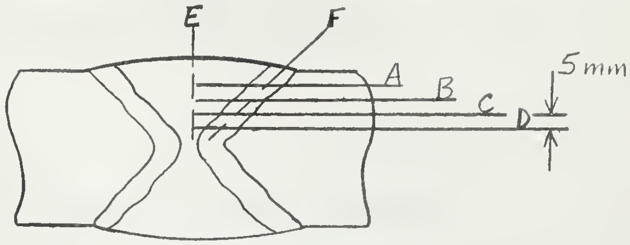
LOCATION OF FRACTURE	CHARPY V-NOTCH FRACTURE ENERGY FT-LBS (at 72 degrees Fahrenheit)
BASE METAL, specimen parallel to direction of roll.	114, 110, 118
BASE METAL, specimen transverse to direction of roll.	89, 88, 108
HEAT-AFFECTED ZONE, adjacent outer passes; non-preheated weld.	37, 45, 41
HEAT-AFFECTED ZONE, adjacent outer passes; preheated to 150 F.	58, 66, 68
HEAT-AFFECTED ZONE, adjacent outer passes; preheat 150 F, and tempered at 1150 F.	71, 86, 72
WELD DEPOSIT, specimen longitudinal.	65, 66, 74
WELD DEPOSIT, specimen transverse.	80, 66, 78

APPENDIX I

TABLE 2A

HARDNESS MEASUREMENTS IN NON-PREHEATED WELDED JOINT

Hardness readings were taken along traverse lines designated A through F in sketch at right.



Distance (mm) from center of weld (A,B,C,D) or from upper surface (E,F)	KNOOP HARDNESS NUMBER					
	A	B	C	D	E	F
1	349	320	320	318	329	465
2	335	320	323	286	349	439
3	356	326	279	332	349	465
4	356	326	340	326	349	420
5	356	329	335	306	335	411
6	358	323	329	282	323	431
7	370	305	356	270	323	431
8	352	314	332	256	326	386
9	358	314	459	---	322	420
10	336	314	425	---	311	349
11	345	367	356	---	320	382
12	341	365	269	---	320	420
13	342	370	267	---	310	332
14	475	335	264	---	323	329
15	501	326	267	---	311	394
16	483	284	267	---	311	370
17	470	284	267	---	311	403
18	472	291	271	---	305	326
19	449	279	279	---	---	374
20	342	294	291	---	---	346

TABLE 2B

HARDNESS MEASUREMENTS IN WELDED JOINT PREHEATED TO 150 F

Distance (mm) from center of weld (A,B,C,D) or from upper surface (E,F)	KNOOP HARDNESS NUMBER					
	A	B	C	D	E	F
1	311	336	382	359	335	439
2	320	329	329	326	352	429
3	311	329	336	359	314	431
4	297	323	326	329	342	452
5	308	349	335	374	339	329
6	320	336	334	367	274	411
7	335	345	356	420	294	398
8	326	328	349	320	339	425
9	335	335	349	249	356	444
10	329	420	370	267	299	425
11	328	454	339	---	326	420
12	305	302	267	---	329	431
13	434	281	267	---	345	390
14	439	274	279	---	305	364
15	449	289	275	---	326	---
16	454	---	---	---	339	---
17	403	---	---	---	317	---
18	---	---	---	---	335	---
19	---	---	---	---	---	---
20	---	---	---	---	---	---

Note: Letters A,B,C,D,E,F designate traverse lines as shown on Table 2A.

APPENDIX I

TABLE 3

HARDNESS MEASUREMENTS IN WELDED JOINTS USING TYPES B88 AND 9018 ELECTRODES

LOCATION	MIL B88 ELECTRODE MIG WELD SPECIMEN "N"		9018 ELECTRODE MANUAL ARC WELD SPECIMEN "U.U."	
	KNOOP HARDNESS	COMPARABLE R_c HARDNESS	KNOOP HARDNESS	COMPARABLE R_c HARDNESS
BASE METAL				
Average	278	25	254	21
Measured Max.	297	28	264	22
Measured Min.	264	22	241	Rb98
HEAT-AFFECTED ZONE				
Average	433	43	418	41
Measured Max.	465	45	508	48
Measured Min.	356	36	349	35
WELD DEPOSIT				
Average	304	29	259	22
Measured Max.	320	31	314	30
Measured Min.	284	26	228	Rb95

Notes:
1. All measurements on polished and etched specimens, using Tukon Hardness Tester.
2. Both specimens received 75 degree preheat and interpass temperature.
3. Specimen "N": no cracks. Specimen "U-U": 2 heat-affected zone cracks.

TABLE 4

NOTCH TOUGHNESS OF RE-TEMPERED BASE PLATE

TEMPERING TEMPERATURE (Fahrenheit)	CHARPY V-NOTCH IMPACT ENERGY, FT-LBS	
	PARALLEL TO ROLLING DIRECTION	PERPENDICULAR TO ROLLING DIRECTION
No temper	114, 110	89, 88
400	113, 115	72, 82
600	118, 127	88, 85
800	124, 120	90, 90
1000	127, 129	101, 95

TABLE 5

TENSILE TESTS OF THE HEAT-AFFECTED ZONE

TENSILE SPECIMEN	PREHEATING TEMPERATURE (Fahrenheit)	ELONGATION, INCHES IN 2" *	PERCENT ELONGATION
1	None	.00	.00
2	None	.08	4.00
3	None	.12	6.00
4	150	.265	13.25
5	150	.010	.50
6	150	.220	11.00

*at time of first visible crack in heat affected zone.

APPENDIX I

TABLE 6

NOTCH TOUGHNESS OF STRAIN-AGED
MODIFIED HY-80 BASE PLATE

TEMPERING TEMPERATURE (Fahrenheit)	CHARPY V-NOTCH IMPACT ENERGY, FT-LBS.
No temper	80, 79
500	73, 71
700	64, 74
900	76, 82